SOME ASPECTS OF THE IRON-CARBON DIAGRAM.

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INTRODUCTION.

SOME ASPECTS OF THE IRON CARBON DIAGRAM.

INTRODUCTION.

In the course of the work described in this thesis, the lines of investigation have been so various, the phenomena examined so diverse, and the conclusions arrived at of such varying importance that none but a general title will serve. It is unfortunate that such a vague title should be necessary, but between the alternative of an explicit title, which might on the whole prove too specific, and a comprehensive title which might on the contrary be somewhat indefinite, the balance of consideration seems to favour the latter choice even at the risk of pretention.

The investigations to be described were undertaken in the first place to find what changes, if any, take place during heat treatment, other than those implicit in the accepted equilibrium diagram of the iron carbon system. From time to time in the course of the work certain departures from the original scheme were made. In the main, however, the scheme was adhered to with considerable persistence. Deviations from the lines laid down were only undertaken when the course of the investigation indicated that certain lines might be followed with advantage.

Although it was not always convenient to do so, efforts were made to follow up any significant indication, not always because something conclusive was anticipated, but to avoid that unsatisfactory state of affairs which arises when, at the conclusion of the work, many points remain to be examined. It is not unusual in the literature of metallurgy to find theories which might easily have been proved or disproved by a simple experiment. Yet the experimenter, though aware of this fact, has neglected to perform it. It is, of course,

course, not possible within any appointed time to clear up every doubtful point which suggests itself, but a realisation of the advisability of so doing is conducive to thorough work.

Frequently, almost invariably, the investigations undertaken in the directions suggested by certain results obtained, ended in disappointment. The hopes that particular deductions would receive further substantiation in these suggested directions, were rarely realised. But these temporary set-backs were not entirely unfortunate. Although the results were not as anticipated, their very unexpectedness proved that the expectation had been unjustified, and the deduction which engendered it unsound. Thus many results have been obtained, which, though inappropriate to the main aims of the work, are not irrelevant when the general significance of the whole comes to be considered. These results enter into the general consideration of the work as obstacles to facile speculation. As obstacles which cannot readily be overcome by overlooking, they are there to check imperfectly considered deductions.

The experimental results, which form the primary subject matter of this thesis, vary in accuracy as they do in nature. This variation in accuracy is inseparable from experimental results obtained under varying conditions. As far as possible the relative accuracy of each method of investigation used will be referred to when dealing with it, and the limits set to rational deductions from the results obtained will be indicated.

Sometimes, in the course of a particular section of the work, it became apparent that modification of the experimental procedure and the conditions observed, would increase the precision of the method. Whenever this occurred the modification was introduced despite the fact that it involved scrapping many weeks' work. Thus in the case of the deter-

-mination of the specific volume of quenched carbon steels, These were performed in the first place on specimens heated in coal gas or nitrogen. Indications that this method of heat treatment compared unfavourably with vacuum treatment led to those results being discarded. Again, the specific volume of all the quenched steels was determined on specimens of varying weight. When it became apparent that the absolute weight of each specimen, and the relative weight of one specimen to another were important factors, the whole series of determinations was repeated. Similarly with the Electrode potential and electrical resistance determinations; these also were repeated whenever an improve-ment suggested itself. As the evolution of each of the final experimental methods will be described fully under the proper heading, further developement of this theme may well be left till then.

In a comparatively comprehensive scheme of research it is important that everything which is going forward should be carefully noted. It is not only that which is sought that matters when the work covers a wide field; although work of a lesser scope may frequently be regarded as either positive or negative. To conform to this injunction is no easy matter. The significance of anything observed cannot usually be appreciated except in relation to something already known or sought, so that, whatever is expected will be the more readily noted. Nevertheless, in certain sections of this present work, notably the electrode potential measure-ments where no guide to the probable results was available, it was only by taking careful account of the most insignifi-cant detail that the work was eventually accomplished. Some

of the subsidary facts noted during the work have not so far proved important, but others have been found essential to the scheme. Whether its relation to the general scheme is now apparent or not, every observation made has been recorded.

The primary object of this research was, as already mentioned, to discover what change took place during quenching and tempering. But the secondary object of the work, namely the developement of methods by means of which the primary object might be achieved, has, on the whole, proved of equal, if not of greater importance. In devising these new methods every effort was made to choose the best possible conditions, to use the most suitable apparatus, and to standardise the procedure exactly. It is not always the case that investigators can choose the best possible apparatus. This is unfortunate, for nothing is more dis-heartening to the worker nor more detrimental to his work than apparatus which is not the best. In carrying out the present work everything necessary in the way of apparatus was provided. No restriction was placed upon the amount of new apparatus bought nor existing apparatus monopolised. If at any point in the work the apparatus used has fallen short of perfection, it is due to ignorance of the existence of more suitable apparatus. The selection of the best possible conditions to be observed was, of course, largely arbitrary. Whenever available the guidance of past work was sought, but the choice was finally dictated by consider-ations of a more general nature. Of course, in certain sections of the work no help could be obtained from published work, and in these instances, the conditions were chosen from considerations based on first principles. In some

some respects this practice was observed throughout the entire work even where, as in the case of the electrical resistance measurements, a voluminous literature was available.

The methods, therefore, by which the work was finally accomplished differ in greater or less degree from the methods used previously by other workers. Where this difference exists it is hoped that it constitutes an improvement, for it is felt that considerations of convenience, of expediency, or of some other kind not directly connected with the objective ends desired, have entered into some methods contrived in the past.

In deciding the experimental procedure in this way, no departure from traditional practice has been made: nevertheless some apology for thus resorting to personal freedom of choice is necessary. Different methods of making the same set of measurements, habitually produce different results, and when the data obtained by a number of workers are considered as a whole for the purpose of some general synthesis, the discrepancy which frequently appears does so as an element of unquestionable mischievousness. That branch of metallurgy with which the present thesis is more particularly concerned is so complex of itself, and the introduction of further complications in the shape of varying composition in the specimens used is so unavoidable, that the exact part played by diversification of experimental procedure is not immediately apparent. But this very complexity of the phenomena examined makes it imperative that no further complication be introduced during the examination. Steel will ultimately be revealed as an essentially simple material to which a putative capriciousness has been unjustly assigned. Therefore variations of method of research should not be introduced without due forethought, and for whatever there is of confusion and obscurity in this thesis, full apologies are offered.

It is unfortunate that the less understood a subject is

is the more liberties can be taken with the study of it. In this way, the general body of knowledge on the subject tends to be held and carried forward in a state of confusion and distortion. Unless some standardised experimental methods are insisted on, unless the terminology is periodically examined, defined and sifted, and the data habitually viewed in their true perspective, subject to all the modifications and limitations imposed on them by the mode of their discovery, unless this is done, the confusion and distortion arising from conflicting results will be cumulative. This will be found to be less true the nearer a certain subject approaches to a mathematical nature, and more true the nearer it approximates to a purely terminological science.

Whatever liberties have been taken with the study of the heat treatment of steel during the present research, have been taken in accordance with common practice. So that, although it is felt that this is not what should have been done, in the absence of standard methods it was the best that could be done.

The importance of the personal factor as an element of confusion in the experimental work is of much less magnitude than in the subsequent deductive phase. Whatever the nature of the experimental results - whether false or true - the process of deduction, generalisation, and synthesis of a general theory is substantially the same. It is purely an application of rational thought, and will be the more successful, though not the more spectacular, in proportion as the laws of logical probability are observed. Knowledge of the data pertaining to the problem and of the course of analagous speculations previously carried out, will contribute to the results achieved: but the value of the achievement will in the main be determined by the degree of mental discipline to which the speculator has been subjected in the past.

This phase of the work is purely philosophical, and will be the more effective according to the extent to which the subject matter of all those sciences pertaining to human thought, is borne in mind. In effect, the everyday principles of common sense and matter-of-fact will be more effective in the achievement of tangible ends than the most extensive recourse to esoteric terminology.

Everything that has been observed during the accumulation of experimental data comes into the general scheme of knowledge held by the scientific community as a whole, either through the medium of publication or address. So that between the performance of the work and its presentation for general service, the data pass through a phase of consideration, of turning over and reviewing in the mind of the worker. It is during this phase that the entire mental content of the individual exercises a pronounced effect in shaping the final form in which the work will be presented, in selecting which results will be held as significant and which will be regarded as neglible variants, and in deciding what conclusions may be drawn.

This aspect of research and of science generally, has received scant attention from scientists as a whole. But gradually, as modern views on the mental processes of mankind become more widely apprehended, it is being realised that the investigator presents a more intractable problem than the phenomena he investigates. Due to this intermingling of personality with objective facts, there is much in the literature of metallurgy which would lend itself more readily to psychological interpretation than to any interpretation in accordance with material facts.

With regard to what has just been said, no special claim of immunity from common weaknesses in this direction is made for the present thesis. It is hoped, however, that the text

text will be found to be comparatively free from incursions into the realm of pure sophistry, and from metaphysics masquerading as tangible facts.

During the work a certain amount of microscopical examination was performed, but it cannot be claimed that this examination was systematic or extensive. Therefore the work, as a whole cannot readily be elucudated in terms of those structures which form the basis of the physical investigation of steel. Despite this, it is essential that the names used to designate these familiar structures be used frequently in the text, if the work is to enter into the general scheme of knowledge of this subject. These names will be used then as distinguishing names for certain phases which appear without reference to the characteristic structures to which they give rise when viewed under the microscope. There is, of course, nothing original in this use of the nomenclature of metallography. Still it is advisable to make clear in which sense use is made of these terms which were originally applied to constituents identified microscopically.

Now that the question of nomenclature has arisen, it may be as well to state what certain terms used in this report are intended to convey. As most of these terms are used in their accepted sense, some justification for the redundancy introduced by restating their meaning is necessary. This justification is found, not in any discrepancy between the prevalent apprehension of these terms and the use made of them here, but in the lax everyday use of them. For example, martensite is generally believed to be a solid solution of carbon in alpha iron. By means of a defective syllogism, some metallurgists have concluded that a solution of carbon in alpha iron is martensite. An ambiguous significance thus becomes attached to the word martensite, and as a result some controversy has arisen as to whether, at the normal change

change point, austenite can break down only through martensite, or direct to pearlite. No one would be so rash as to assert that the constituent martensite, which is microscopically distinguishable, and characterised by extreme hardness and low density, appears as a transition between austenite and pearlite during the normal change. Yet that is what is implied when this use is made of the word martensite.

In this thesis the common metallographic terms will be used in the following sense:

The terms ferrite, cementite, pearlite, sorbite, troostite, will be used in the accepted sense, except where some qualifying statement accompanies the term. It may be advisable to add that troostite, sorbite and pearlite are considered as simply particular distinctive structures in a continuous gradation. Martensite will be used to designate a particular solid solution of carbon in alpha iron, characterised by certain physical properties and by the mode of its formation. This particular solid solution will be considered as perfectly distinguishable by these characteristics from any other solution of carbon in alpha iron. The term austenite will be used in its usual sense as signifying the solid solution formed by carbon and gamma iron. These two solid solutions are distinguishable by their properties from each other and from all other constituents occurring in steel, and the terms can therefore be used without reference to their microstructure. The other constituents, pearlite, sorbite, troostite and ferrite are not so readily distinguishable, and consequently will not be referred to except in terms of their microstructure. For reasons which will be given later, carbon instead of carbide will in all cases be referred to as the solute in the solid solutions.

The work described in this thesis is part of a scheme of research performed jointly by Mr. M.S. Fisher, B.Sc. A.R.T.C. and the writer. When broken into two separate parts the coherence of the whole is lost, and constant reference will therefore be made to the results obtained by Mr. M.S. Fisher. Sometimes these results may be used unconsciously, but so far as possible, whenever a point raised by Mr. Fisher's results comes into the thesis, mention of its source will be made.

In making the electrical resistance determinations and in measuring the continuous change of resistance during tempering, two observers were required. These results are therefore described in the thesis by Mr. M.S. Fisher as well as in the present thesis.

SPECIMENS.

Five series of steels, the analysis of which are tabulated in tables 1 and 2, were examined. Specimens of a shape and size most suitable for the purpose were cut from the bars as required. To obviate any complications arising from different previous heat-treatment, each specimen was used once only. The same specimen was used of course throughout a series of tempering experiments, but at the conclusion of these experiments it was not again quenched for another purpose.

HEAT TREATMENT.

Some of the earlier experiments were performed on specimens heated in coal gas or nitrogen, but the results obtained in this way were ultimately discarded and all subsequent heat-treatment performed in vacuo. The furnace used was of the resistance type, and consisted of a transparent silica tube wound with platinum wire. Using this transparent tube no difficulty was experienced in maintaining a vacuum even at the highest temperature used. A Toepler mercury pump was used to evacuate the tube.

The specimens were heat-treated in sets of five in the case of the nickel and the chromium steels, and of seven in the case of the carbon steels. To ensure equal quenching the specimens were heated in a porcelain boat, to one end of which a looped wire was attached. When ready for quenching the vacuum was broken, the stopper carrying the couple was withdrawn, and by means of the attached wire the boat was pulled into a large volume of iced brine. The vessel containing the iced brine was placed so that the surface of the brine was within a few inches of the end of the furnace tube. The time occupied in transferring the specimens from the tube to the quenching bath was therefore about half a second.

Annealed specimens, cooled slowly in the furnace, were found to be quite untarnished at the conclusion of the heat-treatment. Quenched specimens were slightly oxidised during the time which elapsed between breaking the vacuum and drawing them into the bath.

Temperature was measured by means of platinum/platinum-iridium thermo couples and a Siemens' direct reading galvanometer. The couples were standardised from time to time, against known melting points.

SECTION 1. SUB-SECTION A.

SPECIFIC VOLUME DETERMINATIONS: EXPERIMENTAL.

THE SPECIFIC VOLUME OF STEELS.

In this section, measurements of the specific volume of the steels in the carbon, nickel and chromium series are described. The determinations were made on the steels as quenched from different temperatures, and as annealed. Tempering experiments were carried out with the chromium steels only, and the effect of immersion in liquid air was investigated in the case of carbon and chromium steels.

PRELIMINARY REMARKS.

Certain aspects of the effect of heat-treatment may easily be examined by determining the specific volume. The measurements can be carried out with great accuracy if reasonable care is taken, and if the possible sources of error are constantly borne in mind. No elaborate apparatus is required, but a good balance is indispensable.

These determinations, like all other physical measurements, have only a limited usefulness. Martensite and austenite have very distinct specific volumes, so that the method is preeminently suitable for detecting the presence of either of these constituents. Graphite or temper carbon has also a distinctive specific volume and the presence of even a small amount may be detected. The specific volume of an annealed steel in which some of the carbon is graphitised, is very similar to that of a quenched steel containing martensite, but as a rule knowledge of the previous heat-treatment, and in doubtful circumstances microscopical examination, will avoid any confusion on this account.

Benedicks (2) has shown that sorbite, troostite and pearlite all have the same specific volume. There is much in the present work that goes to show that this statement cannot be accepted without qualification. The point will, however, be raised again. The purpose of these preliminary remarks is

is only to establish a few definite points which will be a guide to the interpretation of the results obtained.

The statement made above with regard to the specific volume of martensite and austenite, will also require modification. Neither of these constituents possesses a definite invariable composition. In fact the composition of both will vary within certain fairly wide limits. As the composition varies so will the properties vary, notably the specific volume. Recent work on the X-ray spectrum of steel has made it abundantly clear that, as the carbon contained in solution in alpha iron or gamma iron is increased, so will the lattice be expanded and the specific volume increased. This variation complicates the results a little and makes quantitative interpretation difficult.

Sufficient has been said to indicate the main factors which arise in considering specific volume results.

That the measurement of specific volume, despite its convenience, has hitherto been seldom used is testified by the meagre data available. Reference to past work will be made in considering the results obtained during the present investigation.

METHOD OF DETERMINATION.

The method employed for the specific volume measurements was based on that described by Andrew and Honeyman (1). As some slight modification in the procedure has been introduced, the method will be described briefly.

The specific gravity of a quantity of pure paraffin oil was measured at 15° C. by means of a specific gravity bottle. By repeated weighings in air and oil, the specific volume at 15° C. of an annealed carbon steel was accurately determined. Throughout the experiments this specimen was used as a standard.

During the weighing of a series of specimens in oil the

the temperature increased gradually. The standard specimen was weighed before and after each specimen, and if the two values of the standard differed by more than .2 milligrammes, due to alteration of the temperature of the oil, the weighings were repeated.

The weights used were standardised occasionally during the experiments. Repeated determination of the specific volume of the same specimen showed that the possible error was less than ± .00001.

All specimens used in the final recorded determinations were in the form of thin sections weighing 18 grammes. It was found that with a smaller weight the method lost something of its accuracy, while if the weight was increased the attendant increase in size produced inefficient quenching. Considering the process as a whole - heat treatment and specific volume measurement - the exact size and shape of the specimens used are therefore important. This, of course, applies with more force to the relative weight and shape of the specimens in a series, than it does in the absolute sense.

Except during the tempering experiments and the liquid air treatment, a new specimen was used for each determination. Before quenching, the specimens were soaked at the quenching temperature for 1 hour. Annealing was accompanied by soaking for 3 hours at 850° and cooling in the furnace. In the tempering experiments a glycerine bath was used at the lower temperature, but above 200° the tempering was carried out in the electric furnace.

During the calculation of the specific volume of the specimens from the weight in air and in oil of known specific gravity, seven figure logarithms were used.

The results obtained are tabulated in tables 4 and 5, and plotted in figures 1 to 5. No correction for manganese phosphorous or silicon has been made to the results recorded,

recorded, but the factors, by means of which the correction may be made and the requisite correction for each steel, are given in table 3. The results have not been corrected because it is felt that the effect of a particular element cannot readily be expressed as a factor applicable in all cases, and because the relation between the results when plotted is not improved by correction.

SECTION 1. SUB-SECTION B.

SPECIFIC VOLUME DETERMINATIONS: CONSIDERATION OF RESULTS.

SPECIFIC VOLUME. CONSIDERATION OF RESULTS.

The specific volume determinations are tabulated in Tables 4, 5, and 12, and plotted in Figures 1 to 10. As the results are somewhat diverse they cannot be easily discussed as a whole. The figures are designed to facilitate comparison of the results and to emphasise certain points. For clearness and for convenience, the results will be considered as they appear in the figures.

In Figures 1 and 2 the specific volume measurements, made on chromium steels after quenching from 1100° C. and after tempering at progressively rising temperatures, are shown. Both figures contain the same determinations. In Figure 1 they are plotted against carbon content, and each line represents the specific volume of the series after 3 hours at the tempering temperature, denoted by the degrees centigrade at end of line. In Figure 2 the specific volume results are plotted against tempering temperature, and each line represents a steel in the chromium series.

The most interesting results are those obtained with the chromium steel B (1.73 per cent carbon). When quenched, this steel contained a notable amount of austenite, and the results obtained by tempering it record some interesting features of the tempering of this constituent. Other steels in the series contained austenite when quenched, but martensite was the predominant constituent. The steel with 1.73 per cent of carbon contained very little martensite.

Tempering at 100° C. for 3 hours produced a decrease in the specific volume of all the steels. This decrease must be attributed to decomposition of martensite, which has a much higher specific volume than the troostite formed from it. In each case the change in specific volume should be proportional to the amount of martensite in the quenched steel. From this

this it may be deduced that there is a maximum amount of martensite in the steel containing .98 per cent of carbon, and that the amount decreases rapidly as the content of carbon is increased or decreased.

The phrase "amount of martensite" is not used quite accurately in the foregoing. In the quenched state the structure of all the steels with less than 1 per cent of carbon was typically martensitic when examined under the microscope. Therefore in the original metallographic sense they all consisted entirely of martensite. The word martensite as used in the present connection has no relation to microscopic appearance, but means simply a solid solution of carbon in alpha iron. As the carbon content of the steel is raised the amount of carbon dissolved in the alpha iron increases. This increase is evidently accompanied by an expansion of the alpha iron lattice which results in a rise in specific volume. Thus all the steels between .35 and 1.0 per cent of carbon are entirely martensitic under the microscope, and they are all composed of a solid solution of carbon in alpha iron; but the concentration of carbon in this solid solution increases with the carbon content of the steel, and as this increased concentration of carbon is accompanied by a distention of the lattice, the specific volume increases with carbon content until austenite appears.

Within a certain range of composition-in the steels under discussion between .8 and 1.0 per cent of carbon - the decrease in the specific volume of quenched specimens caused by austenite, is not sufficient to balance the rise caused by increased concentration of carbon in the martensite. Within this range, although some austenite is formed, the specific volume continues to rise as the carbon content of the steel is increased, but at a lesser rate than before. Therefore the specific volume curve bends off about .7 per cent carbon.

The pronounced change in direction of the curve for

for quenched steels at .98 per cent of carbon is therefore incorrect. Probably the change should take place gradually, but in the absence of a sufficient number of steels in the neighbourhood of this composition, the curve cannot be drawn otherwise. It should be noted, however, that austenite does not appear suddenly in large quantities once a particular composition is reached, but gradually and in increasing amounts as the carbon content rises.

These results do not show clearly whether given increases in the carbon content are always accompanied by the same increase in the amount of austenite, or whether the amount of austenite retained is other than a linear function of the carbon content. The amount of austenite, which at first increases gradually as the carbon content rises, might, after a particular composition is passed, increase more rapidly.

Another point on which the results are not clear is the mechanism of the tempering of martensite - or rather of the deposition of carbon from solution in alpha iron. They show that there is a decrease in specific volume at 100° C. and 150° C., but they do not show directly whether this is caused by a separation of some carbon from the whole solution or by the complete decomposition of certain parts. This point is important and is not irrelevant.

In Figures 1 and 2 it will be seen that the steels containing between .62 and .98 per cent of carbon contract considerably on tempering at 100° C., whereas those with less than this amount of carbon contract very slightly at that temperature. It will also be seen that after tempering at 150° C. all steels between .62 and .98 per cent of carbon have the same specific volume.

At 200° C. Austenite begins to temper and causes a slight rise in the specific volume of the steel containing 1.73 per cent of carbon. The steels with .8 and .98 per cent carbon continue

continue to contract, but not so much as before; austenite is apparently tempering but the contraction of the alpha solid solution is still the predominant effect. The most noticeable feature of the tempering at 200°C. is the appreciable change in the specific volume of the steel containing .62 per cent of carbon. Tempering at lower temperatures produced no great change in the specific volume of this steel.

At 250°C. the effect of the tempering of austenite is very marked in all steels above .62 per cent carbon. The specific volume of the steel containing .62 per cent carbon undergoes little change at this temperature, and it must be assumed that in this steel the effect of the expansion caused by decomposition of austenite is about balanced by the contraction caused by the tempering of martensite.

Tempering at 300°C. produces a further marked contraction of the steel containing .62 per cent of carbon.

Tempering at 350°C. produces the greatest change in the specific volume of the steel containing .35 per cent of carbon.

The foregoing remarks on the tempering of martensite, as observed by specific volume measurement, may now be summarised. It is evident that the higher the carbon concentration in the alpha solid solution, the lower the temperature at which the greatest contraction occurs. Thus the steel containing .8 and .98 per cent of carbon contract considerably at 100°C., that containing .62 per cent contracts considerably at 200°C., and that containing .35 per cent contracts considerably at 350°C. These results have nothing to do with austenite. Any tempering of this constituent at low temperatures will tend to minimise rather than cause the effect noted. The effect is due to the tempering of the alpha solid solution only.

Having remarked on the principal features of the curves so far as they refer to the decomposition of the alpha solid solution, discussion of their bearing on the mechanism of this

this change may be resumed.

The amount of change which takes place in this alpha solid solution during tempering, may be related in various ways to the carbon content and tempering temperature. For instance, the amount of change may be proportional to the tempering temperature and time of tempering. In this case tempering at a given temperature for a given time would produce an identical change in every steel in the series. The steel with the lowest carbon content would reach its stable annealed value before any other steel in the series. Time and temperature would be interchangeable.

This conception of the process of tempering of martensite is widely held, though rarely explained or defined. These specific volume results show that the amount of change is not related so simply to time and tempering temperature.

The amount of change which takes place during a particular time at a given temperature may be proportional to the carbon content. In this case the highest carbon steels would change most, and the lowest carbon steels would change least during tempering for a similar time at a given temperature. In another section it will be shown that during tempering at 240° C. these conditions are approximately true. When the steels as quenched were suddenly plunged into an oil bath at constant temperature, the rate of decomposition of martensite was proportional to the carbon content.

As the specific volume results show, these conditions do not hold during tempering at progressively rising temperatures. In this case the higher the carbon content of the alpha solid solution, the lower the temperature at which rapid decomposition can commence. At temperatures higher that 300° C. the amount of change is roughly proportional to the carbon content.

These two sets of results, the specific volume under discussion and the tempering resistance curves, section 4, are not

not contradictory. Taken together, they show that the more concentrated the alpha solid solution is, the lower the temperature at which it decomposes, but when tempering is performed at a temperature higher than the critical tempering temperature of all the steels, the amount of change in a given time is proportional to the carbon content.

Because of the interference of austenitic tempering in some of the steels, the specific volume results do not prove the foregoing idea, but they suggest that something of that nature is occurring.

It has already been stated that these results do not directly explain the mechanism of the decomposition of the alpha solid solution. They suggest, however, that there are some aspects of the process which are not yet fully understood. If a uniform solid solution of carbon in alpha iron decomposes by gradual separation of carbon from the whole mass of this solution, and if the most concentrated solutions decompose at the lowest temperatures, then when the concentration of carbon in the uniform solid solution has reached a critical amount no further decomposition will take place at that particular temperature.

But microscopic examination does not support the view that the decomposition of martensite proceeds uniformly at low temperatures. The evidence rather suggests that decomposition of the solid solution takes place with production of troostite in certain areas. If this is what actually occurs it is difficult to see why the process, once commenced, does not proceed to completion at the rate at which it started. If one area can decompose, why do other areas remain unchanged?

There is no doubt that tempering proceeds slowly for an indefinite time at low temperatures, but the rate of change is vastly different from that at which the process commenced. This diminution of the rate of decomposition can be explained by assuming that the solid solution is separating carbon uniformly,

uniformly, and that the rate of separation is decreasing as the process proceeds. There is no other evidence in favour of this view, and as the sudden decomposition of certain areas in a uniform whole cannot account for the falling off and ultimate cessation of tempering before the reaction is complete, some other factor must influence the process. What the other factor is, the results do not indicate, but they make it clear that martensite is not a uniform solid solution of carbon in alpha iron.

Specific volume determinations are very convenient for separating the tempering of austenite and martensite. The separation of course, is not perfect, for the method takes account of the total volume changes in a specimen and therefore registers the dominant effect, whether expansion or contraction. Many methods of research possess this disadvantage. It cannot therefore be said that there is no austenitic tempering below 200°C. A little of this constituent might temper and the expansion accompanying the change be obscured by the contraction of martensite. At 200°C. the effect of the tempering of austenite is just noticeable, but at 250°C. it is considerable. This tempering of austenite causes a rise in the specific volume of the three steels containing .8, .98, and 1.73 per cent of carbon respectively, and appears to balance the martensitic contraction in the steel containing .62 per cent of carbon.

Tempering at 300°C. produces a further rise in the specific volume of the steel containing 1.73 per cent of carbon, showing that more austenite has tempered. Thereafter as tempering is continued, the specific volume of this steel decreases due to the tempering of the martensite formed from austenite.

Similar results have been obtained by a number of investigators. Maurer (3) noted a large increase in the specific volume of a 1.66 per cent carbon steel, quenched from 1050°C. and tempered at 200°C and 250°C. Enlund (4) obtained evidence

evidence of this formation of martensite from austenite when making determinations of the specific volume of a number of carbon steels. And Andrew and Honeyman (1) obtained expansion on tempering quenched carbon steels at 250°C.

The work of all these investigators shows than on tempering austenite in the neighbourhood of 250°C. it changes to martensite, the change being accompanied by an increase in volume. The results obtained in the present investigation confirm this.

These results, although apparently straightforward, are really difficult to explain. If martensite tempers more readily and rapidly than austenite, as the tempering at low temperatures indicates, how does martensite accumulate from the tempering of austenite at a particular temperature? It would appear that in the neighbourhood of 250°C., the relative rates of tempering of these two constituents are reversed, so that austenite tempers more rapidly than martensite. This is one way in which the results can be explained.

In section 4 it will be shown that at 240°C. austenite tempers much more slowly than martensite, so that this way of explaining the observed facts is untenable. This matter is discussed more fully in section 4, but it may be said here that the only possible explanation of the facts is than martensite does not form from austenite at the tempering temperature, but during cooling from the tempering temperature. This brings the formation of martensite from austenite into line with the the formation of martensite during quenching.

Besides overcoming the difficulties raised by the specific volume results, this idea explains why the expansion accompanying the change from austenite to martensite, has never been noted in dilatation curves of austenitic and semi-austenitic steels. By showing the similarity of the two ways of producing martensite - by quenching and by tempering - this conception of the tempering process assists in clearing up difficulties which have their

their origin in the belief that these two ways are essentially different. For instance, in the discussion of Hallimond's(5) paper "On Delayed Crystallization in the Carbon Steels: the formation of pearlite, troostite, and martensite." Rosenhain said, "It had to be remembered that the formation of martensite in ordinary carbon steels during quenching was really a very exceptional phenomenon, and was by no means the only way in which martensite ------ could arise. Certain alloy steels when quenched were soft. An alloy steel could be taken and quenched and then a hardened martensitic condition obtained by subsequent reheating. Doubtless the theory of supersaturation would lend itself also to the explanation of that phenomenon------"

These comments on Hallimond's theory of supersaturation reflect current opinion on the formation of martensite. Obviously, if the formation of martensite during tempering and during quenching takes place in an identical way, the theory of supersaturation need not be extended to cover the tempering process: it does so without alteration.

In Figures 3, 4 and 5 are shown the specific volume measurements made on the steels of the shromium, carbon and nickel series in the annealed condition, and after quenching from various temperatures.

From specific volume measurements made on groups of non-ferrous alloys the following laws have been established:-

(1) The specific volume of mixtures of two components is a linear function of the composition and can be calculated from the specific volume of the single components.

(2) The specific volume of a continuous series of solid solutions is a linear function of the composition. In certain solid solutions and eutectic mixtures small deviations from the straight line have been observed.

(3) The formation of a compound is indicated by a break in specific volume curves.

The lines representing the specific volume of annealed steels should be straight lines corresponding to varying mixtures of ferrite and cementite. In Figure 4 the line representing the specific volume of annealed carbon steels is practically straight above .48 per cent of carbon, but the annealed lines in Figures 3 and 5 do not correspond so closely with the mixture law. The specific volume of annealed chromium steels deviates considerably from a straight line, and it is evident that, until the eutectoid composition is reached, the constitution of annealed chromium steels is not a simple mixture of varying amounts of ferrite and pearlite. One of the constituents is apparently altering as the carbon content is raised. Probably the ferrite is retaining increasing amounts of carbon in solution. This would account for the more rapid rise in specific volume below .9 per cent of carbon, and as this effect is greater in the chromium than in the carbon steels, it indicates that chromium increases the solubility of carbon in ferrite. Of course, the presence of chromium in solution would increase the specific volume of the ferrite, but there is no reason to suppose that the amount of chromium in solution should increase with the carbon content. It may be that this rapid rise in the specific volume is not caused by alteration in the constitution of the ferrite, but by a change in the composition of the carbide. Moissan (6) prepared in the electric furnace . the chromium carbides Cr_3C_2 and Cr_4C and determined their specific volumes as .15450 and .1481 respectively. Andrew and Honeyman determined the specific volume of Fe_3C Cr_3C_2 as .1438. Various determinations of the specific volume of Fe_3C have been made, and the following values may be taken as representative.

Moissan	.1414
Benedicks	.1292
Levin and Dornhecker	.1317
Westgren and Pragmen	.1312
Andrew and Honeyman	.1306

From the foregoing it is evident that chromium carbide possesses a higher specific volume than iron carbide. Therefore an increase in the amount of chromium would increase the specific volume of the steel. But before suggesting that this is a possible explanation of the rapid increase with carbon content of the specific volume of annealed chromium steels, some account must be taken of what this idea implies.

In these steels the chromium content is constant while the carbon varies. If therefore, more chromium carbide is formed as the carbon content is increased from .35 to .98 per cent, less chromium must remain in solution. Whether the specific volume of the steel will increase as more chromium carbide is formed, depends on the relative effect on specific volume of chromium in solution in ferrite and of chromium carbide.

It has been stated that in annealed chromium steels no free cementite is formed until all the chromium is used up in the formation of the double carbide $Fe_3C\ Cr_3C_2$. If this double carbide is really more than a momentary relationship between the carbides, it must be assumed that as long as there is chromium in excess of that required for the double carbide, the carbon will be distributed between iron and chromium in the proportions represented by the formula. The specific volume should vary regularly as the carbon content is increased, until all the chromium in the steel has combined with carbon. Thereafter, as the carbon content is increased, only iron carbide should form, and the specific volume should increase less rapidly because of the lower specific volume of this carbide. In these steels with a constant chromium content of approximately 1.65 per cent, all the chromium should be in combination with carbon at .38 per cent of carbon. Therefore the rapid increase of specific volume between .35 and .98 per cent of carbon is not caused by an increase in the amount of chromium carbide, unless the proportion of chromium in the carbide is much smaller than is represented by

by the formula Fe_3C Cr_3C_2. Most of the other carbides that have been suggested contain an even greater proportion of chromium than the double carbide, and as a certain amount of chromium in solution probably raises the specific volume more than an equal amount as carbide, it may be concluded that the facts under discussion cannot be explained by assuming the formation of chromium carbide. It is evident that the facts are best explained by supposing that the presence of chromium increases the amount of carbon retained in solution by the ferrite of slowly cooled steels.

In Figure 4 the line representing the specific volume of annealed carbon steels is compared with the mixture line for iron and cementite. In calculating this mixture line, the specific volumes of iron and cementite were taken as .12736 and .1312 as determined by Benedicks and Westgren and Pragmen respectively. The agreement is very good, and although the results obtained in the present investigation indicate that the specific volume of iron should be a little lower and that of cementite a little higher than was found by the above authors, the variation of specific volume with carbon content is obviously in substantial agreement with the mixture law.

The effect of quenching from different temperatures may now be considered.

In Figure 3 the effect of the quenching temperature on the specific volume of chromium steels is shown. It is clear, from comparison of the line representing the specific volume after quenching from 800°C. with that representing the specific volume after quenching from 900° C., that this change in the quenching temperature has produced an appreciable alteration in the state of expansion of the martensite or alpha solid solution. A more detailed comparison of these two curves shows that the magnitude of this alteration in the specific volume varies with the carbon content. It is greatest in the steels of intermediate carbon content, that is, those containing .80 and .98 per cent of carbon

carbon respectively. In the steel containing 1.73 per cent of carbon, only a portion of the carbon would be in solution at 900° C., and this steel, when quenched from 900° C., has a specific volume not much higher than that of the steel containing .98 per cent. Yet this is the highest specific volume obtained for the steel with 1.73 per cent of carbon, for quenching from 1000° C. and 1100° C. results in the retention of austenite. Perhaps austenite is retained on quenching from 900° C., but it cannot be deduced from these quenching results alone whether this is so.

Raising the quenching temperature from 900° C. to 1000° C. produces a further rise in the specific volume of the steels of intermediate carbon content, but raising the quenching temperature to 1100° C. does not produce any further change. The tempering results show that these intermediate steels contain austenite when quenched from 1100° C., and it is probable that quenching from 1000° C. has a similar effect. It must be assumed that the decrease in specific volume, which ordinarily should accompany the retention of austenite, is, in this instance balanced by the increase produced by the further distention of the martensite. It is probably correct to suppose that if no austenite were formed, the specific volume of all the steels in the series would lie on the line obtained by producing the first steep portion of the curve (between .35 and .62 per cent carbon). This would only apply to quenching from temperatures at which all the carbon in every steel in the series was wholly in solution. In Figure 6 the effect of austenite on the specific volume is shown. It is clear that austenite must have a very low specific volume. The exact value cannot be calculated from the present results, as the specific volume of austenite as well as the respective amounts of austenite and martensite, varies with the carbon content, and unless one of these variables is known, the other cannot be found.

In Figure 4 the specific volumes of plain carbon steels quenched from different temperatures are shown. What has been said about the effect of raising the quenching temperature on

on chromium steels is equally applicable to the carbon steels. In these steels, however, the amount of austenite produced in the high carbon members of the series by quenching from 1000°C. and 1100°C., is considerably less than in similar chromium steels. The presence of chromium clearly assists in preventing the decomposition of austenite during cooling.

Besides the curves representing the effect of quenching from different temperatures, Figure 4 also contains three curves, A, B and C, showing the effect of heating to 1100°C. and cooling to the quenching temperature. The specific volume of the high carbon steels quenched in this way is considerably higher than that of the same steels quenched in the ordinary manner. These experiments were carried out to show that the two methods of reaching the quenching temperature do not give identical results. Saldau (7), when quenching specimens for electrical resistance determinations heated them to 1100°C., cooled to the quenching temperature, soaked for ten minutes and quenched. The general form of the curves he obtained for a series of carbon steels quenched in this way from different temperatures, is somewhat different from the curves obtained during the present investigation. It was concluded from this that the two methods of reaching the quenching temperature must give rise to different results. It was argued that on account of the comparatively slow rate of diffusion of carbon, the amount in solution in hyper-eutectoid steels at a particular temperature, would be greater when the steel was cooled to the temperature and less when it was heated to it. And further, during heating the distribution of carbon would be less uniform then during cooling, so that, in steels which had been heated to the quenching temperature and no higher, there would be more likelihood of the existence of areas of high carbon concentration liable to produce austenite on quenching. It was therefore expected that heating to 1100°C. and quenching from a lower temperature, would produce a structure differing from that produced

produced by ordinary quenching from the lower temperature. The specific volume results obtained by quenching in this way are in accordance with the arguments set forth above, but microscopic examination of the specimens which gave these results revealed the presence of graphite. As graphite itself has a high specific volume, its presence in these specimens would no doubt account for some of the observed increase, and as the exact magnitude of the effect of this graphite is not known, it is impossible to discover what portion of the increase should properly be attributed to other differences in constitution. The idea that carbon exerts a similar effect on the specific volume of steel, whether it is present as graphite or in solution in alpha iron, will presently be advanced, but in the results under discussion it is the conversion of cementite to graphite that must be taken account of. Steels cooled from 1100° C. and quenched from 800° C. should normally contain some free cementite, and if graphite instead of cementite separates, the specific volume will be considerably increased.

According to specific volume determinations made on non-ferrous alloys, it appears that in a series of solid solutions formed by any two metals, the specific volume varies with the composition in the same way as when the metals form a mixture.

Bragg (8) has pointed out that if the atoms of a lattice are considered to be surrounded by spherical regions of influence touching each other, an approximately constant radius of influence may be ascribed to each kind of atom. If the foregoing assumption is correct, the space occupied by an atom of a certain kind will be a constant property of the atom, whether the lattice in which it exists is composed entirely of similar atoms or not. Vegard (9) has shown that when atoms of related nature replace one another in solid solutions, so that the two kinds of atoms have a common lattice, the lattice seems to be built of atoms of a unit kind having a radius between those of the interchangeable atoms. Bain (10) has confirmed this for a number of binary alloys.

All this points to the same conclusion, that the specific volume of a metal containing two components in definite proportions, should be the same whether the components form a mixture or a solid solution.

It has been shown that the specific volume of carbon steels conforms closely to the line for mixtures or iron and cementite. It will be seen in Figures 3, 4 and 5, that if the annealed line approximates to the mixture line for each series of steels, the quenched steels deviate considerably from it. Martensite, if it is a solid solution of cementite in alpha iron, has evidently a much higher specific volume than would be obtained by calculation from the known specific volume of its components.

There is a considerable amount of evidence in favour of the view that cementite does not exist in solid solution as such, but breaks down to iron and carbon. Jeffries and Archer have stated their reasons for believing that carbon is in solution as carbon atoms, and not as molecules of carbide. The substitution theory of solid solutions suggested by Rosenhain (11) practically demands acceptance of this idea.

Westgren (12) explains the effect of dissolved carbon on the lattice parameters of austenite, by assuming that carbon atoms alone migrate in the solution. And many others contend that the idea of complex molecules existing in solid solution in a metal, is inconsistent with the modern conception of the solid state.

It is therefore interesting to compare the specific volume of martensitic steels with the specific volume of mixtures of iron and graphite. In Figure 4, line G represents the specific volume, calculated from the mixture law, of a series of mixtures of graphite and alpha iron. In calculating this line, the specific volume of alpha iron was obtained by extrapolating the line representing the specific volume of annealed steels. The specific volume of graphite was taken as .43480, the figure

figure commonly given in physical tables. The slope of this line is similar to the slope of the martensitic line. The lines certainly do not coincide, but when the scale on which they are plotted is taken into account, the absolute value of the difference is evidently not great.

The foregoing can hardly be accepted as conclusive proof of the nature of martensite, but it certainly lends support to the view set forth above. As the matter is of considerable importance further discussion may not be unjustified. This discussion must centre round the relative effect on the specific volume of carbon in solution in alpha iron, and carbon existing as free graphite.

If martensite is a solid solution of atomic carbon in alpha iron, and if the specific volume of the solid solutions produced by varying amounts of carbon are calculable from the mixture law, then, as the specific volume of mixtures of iron and graphite are calculable in this way, the effect of carbon on the specific volume should be the same whether in solution as atomic carbon or out of solution as graphite.

The points X and Y in Figure 4 were determined by Mr. M.S. Fisher. They represent the specific volume of slowly cooled steels containing 1.61 and 1.73 per cent of carbon, in which the greater part of the pro-eutectoid cementite had been converted to graphite. Assuming that all the carbon above .9 per cent existed as graphite in these steels, it can be calculated that the conversion to graphite of .83 per cent of carbon, previously existing as cementite, produces an increase of .00325 in the specific volume, whereas a similar amount of carbon going into solution produces, according to the martensitic line (D in Fig. 4), an increase of .00200 in the specific volume. Thus it appears that the conversion of cementite to graphite causes a greater increase in the specific volume than is produced by the solution of a similar amount of carbon in alpha iron. Before accepting

accepting this as conclusive proof that there is no close relation between the specific volumes of solutions of carbon in alpha iron and mixtures of iron and graphite, account may be taken of the effect of small amounts of austenite on the alope of the martensite line D. If any austenite is retained in the low carbon steels when quenched, the straight line D through the specific volume of these steels, does not truly represent the effect of dissolved carbon on the specific volume of alpha iron.

The specific volume determinations made on tempered chrome steels indicate that even in the steel with .62 per cent of carbon, sufficient austenite was retained by quenching to cause a break in the curve of specific volume against tempering temperature (figure 2). The results obtained by Andrew and Honeyman when measuring the specific volume of tempered carbon steels, indicate that there is austenite in[a] .9 per cent carbon steel when quenched. It is probable that in both carbon and chromium steels some austenite is retained by quenching steels of lower carbon content than those mentioned. The fact, already referred to, that measurement of specific volume takes account of the predominant change only, would explain why no direct evidence was obtained of the presence of austenite in low carbon steels.

The method used by Enlund (4) is much more sensitive to the presence of small quantities of austenite. This author made very accurate determinations of the continuous change of resistance of quenched specimens during progressive heating, and of the change in resistance produced by tempering for certain periods at gradually rising temperatures, followed by cooling to ordinary temperatures. He investigated a series of thirteen steels varying in carbon content from .21 to 1.57, and concluded from his results that "all carbon steels quenched in water in the ordinary way contain some austenite."

If this is so, it explains why the line through the specific volume of quenched steels when produced cuts the specific volume

volume axis at a lower point than the line through the specific volume of annealed steels. It suggests that the effect of dissolved carbon on the specific volume of alpha iron is actually greater than is indicated by the specific volume results obtained in the present investigation. And it supports the view that the true line for a series of solid solutions of carbon in alpha iron, is identical with the line representing the specific volume of mixtures of iron and graphite.

Unfortunately, it is not known how much austenite is retained in the various steels, nor how the specific volume of austenite varies with carbon content. It is therefore a matter of some difficulty to make the argument conclusive by means of a quantitative example.

The only way in which the specific volume of the austenite found in unknown quantities in semi-austenitic steels can be determined, is by means of X-ray studies of the crystal structure. Westgren and Phragmén have measured the lattice parameter of austenite containing .9 and 1.7 per cent of carbon respectively. They found that the lattice parameter was 3.606 Å when the austenite contained .9 per cent of carbon, and 3.629 Å when it contained 1.7 per cent. From their results they concluded that austenite is not formed by simple substitution, but is actually an additive product with the carbon atoms situated in the interstices between the metal atoms. Assuming this to be true, they calculated from their parameter measurement the density of an austenitic manganese steel, and found the result in absolute agreement with the density determined in the usual way.

In the same way it is possible to calculate the specific volume of the austenite containing .9 and 1.7 per cent of carbon.

Assuming, as Westgren and Phragmén do, that the iron atoms alone compose the lattice, the average weight of the atoms in the lattice of the above austenites is $56.350 \times 1.650 \times 10^{-24}$ gr. and $56.805 \times 1.650 \times 10^{-24}$ gr. respectively. The specific volume

volume of austenite containing .9 per cent of carbon is therefore

$$\frac{(3.606)^3 \times 10^{-24}}{4 \times 56.35 \times 1.650 \times 10^{-24}} = .12605$$

taking the parameter measurement already mentioned, and allowing 4 atoms per unit cube in the face centred system.

In the same way the specific volume of austenite containing 1.7 per cent of carbon is

$$\frac{(3.629)^3 \times 10^{-24}}{4 \times 56.805 \times 1.650 \times 10^{-24}} = .12681$$

These values for the specific volume of austenite are plotted in Figure 6A. The straight line through them meets the specific volume axis at .12520, which is therefore the specific volume of pure gamma iron, if these measurements are accurate and the method of calculation correct.

If the mixture line for iron and graphite is considered to be the true line for the specific volume of solid solutions of carbon in alpha iron, it is possible to calculate the amount of austenite in each quenched steel. The line B in Figure 6A shows the percentage of austenite present in water quenched steels of varying carbon content.

It is too much to expect that the results of these calculations should be accurate, but they are probably of the right order.

If the further assumption is made that chromium forms solid solutions with iron, and that these solid solutions have specific volumes in accordance with the mixture law, it can be calculated that austenite containing 1.7 per cent of carbon and 1.65 per cent of chromium will have the specific volume .12720. Chromium steel B which contains these amounts of chromium and carbon, gave the value .12732 when quenched from 1100°C. This means that Chromium steel B was almost entirely austenitic when quenched from 1100° C. This result is confirmed by the small

small change which took place in the specific volume of this steel during tempering at low temperatures (fig.1).

In another section the nature of the solid solutions which appear in steel will be discussed from another standpoint.

Figure 7 shows the effect of prolonged immersion in liquid air on the specific volume of carbon and chromium steels. It is clear from the curves shown that this treatment causes a considerable amount of austenite to change to martensite. A more detailed consideration of these results is rather difficult in view of what has been said about the presence of austenite in all carbon steels. If all steels contain austenite, and austenite is converted to martensite at low temperature, it follows that the specific volume of all steels should show some increase when subjected to this treatment. But the results obtained show that immersion in liquid air has a noticeable effect on the specific volume of high carbon steels only. According to these results, there is no austenite in chromium steels with less than .87 per cent of carbon, nor in carbon steels with less than .9 per cent. Yet the specific volume results obtained by tempering chromium steels at 200° C. and 250° C., indicate the presence of austenite in the steel with .62 per cent of carbon. It is easy to suggest that the austenite present in low carbon steels does not transform to martensite at low temperatures, but it is difficult to discover why this should be so. In the absence of further evidence on this point, it must be admitted that the results obtained by immersion in liquid air support the conclusion that there is no austenite in either chromium or carbon steels of low carbon content.

It is curious that the carbon steels containing 1.61 and 1.73 per cent of carbon should have almost the same specific volume after immersion as the steel containing 1.19 per cent. Perhaps the change from austenite to martensite takes place slowly at low temperatures, and has not proceeded to completion in these steels. Perhaps the martensite formed undergoes a

a change during heating from -190° C. to ordinary temperature.

The chromium steels do not show this peculiarity. The specific volume of all the steels between .62 and 1.73 per cent of carbon lie on a straight line.

It cannot be definitely decided from these results whether immersion in liquid air causes austenite to transform spontaneously to martensite without the accompaniment of secondary reactions. It was expected that this treatment would cause the allotropic change to proceed to completion and give a series of steels consisting entirely of martensite. Although the results show that some change of this nature has occurred, they do not show conclusively that no other change has proceeded simultaneously or subsequently.

In Figures 8, 9 and 10 the specific volumes of the carbon, nickel and chromium steels are compared. The general meaning of these curves is self evident and no comment need be made.

SECTION 2. SUB-SECTION A.

ELECTRODE POTENTIAL DETERMINATIONS: INTRODUCTION.

ELECTRODE POTENTIAL.

In this section, after an introductory survey of past work of a similar nature, a method of measuring the electrode potential of steels is described. The results obtained by applying this method to examination of heat-treated steel are given and their significance discussed.

INTRODUCTORY.

In the course of the many efforts to elucidate the innumerable problems associated with the study of metals and their alloys, recourse has been had to many physical methods of investigation. Some of these physical measurements are more difficult to make than others, and some of the results obtained are difficult to interpret. Nearly every method has its peculiar liability to error. Either there is some intricacy in making the measurements, or some incalculable subsidiary factor which tends to frustrate precise interpretation.

Electrode potential measurement is no exception to this rule. In fact, this method possesses in a peculiar degree those elements which render experimental work laborious and speculation hazardous. For this reason the determination of electrode potential, as a means of investigating the constitution of alloys, has not realised expectations.

The conception of electrolytic solution pressure was introduced by Ostwald, but much of the credit for developing the theory is due to Nernst. He developed the fundamental formula for the electromotive force of a single electrode, and showed that it depends on the ratio of the electrolytic solution pressure to the osmotic pressure of the electrolyte. It is evident that, depending as it does on osmotic pressure, the electromotive force of an electrode will vary with the concentration of the electrolyte. This is so to such an extent that

that there are cells possessing certain electromotive forces dependent only on the different concentrations of the same ion-producing substance. Further, as both the electrolytic solution pressure and osmotic pressure vary with temperature, the electromotive force of a single electrode also varies with temperature.

The electromotive force of a pure metal in solutions of varying concentration, can be determined with comparative ease. But alloys, due to their heterogeneity, have a tendency to react with the electrolyte and so give rise to a change in the surface layer. As only the surface layer is in contact with the electrolyte, it alone influences the electrode potential. Therefore, if the composition of the surface layer changes by the the gradual solution of the most electro-positive metal, the electrode potential of the alloy will also change, as there is no possibility of diffusion and re-establishment of equilibrium within the metal. Eventually the least electro positive of the original constituents of the alloy alone remains on the surface and the alloy shows the potential of this metal.

Alloys which are composed of, or contain heterogeneous mixtures of two or more constituents, present further difficulties due to local electrolytic actions which take place between these microscopic constituents. Thus in steel under certain conditions, the innumerable electrolytic effects taking place between the constituents, may obscure the true electrode potential.

It is customary to measure the electrode potential of a pure metal against a solution of one of its salts, but it is rather difficult to find a suitable electrolyte for alloys. In order that true equilibrium may exist between the alloy and the electrolyte, it is necessary to use a complex electrolyte containing salts of each of the metals forming the alloy, and to vary the proportion of each of these salts according to the relative amounts of the metals in the alloy.

Laurie (13) who performed the first really scientific investigation of the electrode potential of alloys, was not aware of the necessity of providing an appropriate electrolyte. He compared alloys of copper with pure copper in a solution of cuprous iodide, and copper-tin alloys in chloride with pure copper in sulphate. As he took no account of the concentration of the electrolyte, and neglected the conditions necessary to attain proper equilibrium between the alloys and the electrolyte, his values have only a relative meaning. At the time, however, this was quite satisfactory, and Laurie's results served to confirm the existence of the compounds Cu Sn and Cu_2 Sn.

Herschkowitz (14) in his extensive researches on the electrode potential of non-ferrous alloys, made use of a solution of a salt of the most positive metal as an electrolyte. He obtained characteristic curves for a large number of alloys, and observed well marked discontinuities at concentrations corresponding to intermetallic compounds. This method of Herschkowitz has generally been found to give good results, and so long as the components of the alloy occupy widely different positions in the electro-chemical series, it is not open to serious theoretical objection. When the component metals are electro-chemically similar, doubtful results are obtained by using as an electrolyte a simple salt of either.

The true conditions of equilibrium have been investigated by Herschkowitz and Reinders (15) separately. They have established laws which define the exact composition of the electrolyte which will give true equilibrium with a given alloy.

A great number of investigations were carried out by (16) Puskin and his collaborators, who found that the instructions of Reinders were not easy to carry out experimentally. They used acids or alkalies which formed sparingly soluble salts of the most electro-positive metal. When the components of the alloy differed widely in electro-chemical nature, they used a

a solution of a simple salt of the most electro-positive metal.

From what has been said it is clear that in measuring the electrode potential of an alloy three main difficulties are encountered. In the first place there is the liability of the surface layer to change, due to solution of the least noble metal. Further, parasitic electromotive forces between the constituents of a heterogeneous mixture tend to obscure results, and finally a suitable electrolyte is difficult to obtain.

Another point, which has considerable connection with the work described in this thesis, is the variation of electrode potential values with time. Generally, it has been found that the value changes progressively from the moment of first immersion in the electrolyte until a final constant value is reached. There has been some difference of opinion in the past as to which reading, the initial or the final, should be taken as truly characteristic. Most workers have taken the comparatively steady final reading and gradually this has come to be the established practice. For reasons which will be described later, it was found impossible to do this in the case of the steels examined. The variations of the electrode potential were so erratic and of such magnitude that no characteristic reading, other than the initial reading, could be obtained.

Herschkowitz (14) recognises three groups of alloys in which the effect of adding one metal to another produces a certain constitutional effect, accompanied by a characteristic electrode potential change.

In the first group are included all those alloys in which the component metals are mutually insoluble in the solid state, and consequently form mechanical mixtures. In this case the electrode potential is that of the most electro-positive metal. If all the alloys in a series consist of mixtures of the two

two pure metals, the electrode potential of the whole series will be that of the most electro-positive metal. Laurie showed that this law held even when the proportion of the most positive metal was reduced to one part in two thousand. Annealed steels are generally supposed to be mechanical mixtures, but, as will be shown later, they do not conform to the above law. Some allowance must be made for the relatively narrow limits within which the present electrode potential measurements vary. In investigating many of the non-ferrous alloys, the variations found are of such an order, as to render negligible, variations of such magnitude as are found for annealed steels.

In the second group are included all those alloys in which the component metals are soluble in the solid state and form solid solutions. The electrode potential of the solid solution is always less positive than that of the most positive component, and in many cases it is less positive than the least positive component. If two metals form a continuous series of solid solutions the electrode potential varies in a continuous manner through the series.

Intermetallic compounds constitute the third group.

A sharp discontinuity occurs in the potential -concentration curve at compositions corresponding with the composition of compounds. If the compound can exist as such in the solution, and if the solution contains corresponding ions, the compound will have its own electrode potential.

Different allotropic modifications of the same element have also different electrode potentials. This has been experimentally shown. In the case of the gradual appearance of a second modification in alloys which are predominantly composed of the first modification, no evidence of this appearance will be obtained unless the second modification is more electro-positive than the first. When the second modification is more electro

electro-positive than the first, the presence of very minute quantities should be immediately apparent. When, however, the second modification is less electro-positive than the first, its presence will go undetected until the alloy is composed entirely of this second modification. These considerations are of some importance in measuring the electrode potential of steel, especially when a discontinuity occurs at compositions where austenite may reasonably be expected to appear in quenched steels. In the present work such a discontinuity does occur, but it is doubtful whether it is of the requisite nature.

This point does, in fact, give rise to some difficulty in interpretating the results obtained in the present investigation. For although the discontinuity which occurs in the neighbourhood of .9 per cent of carbon is not typical of the appearance of a second modification, it might, under the peculiar circumstances operating in quenched steels, indicate the appearance of austenite in small quantities. At first sight it appears that this point might easily be elucidated, but unfortunately this is not the case. When special elements are added in quantities, sufficient to make the quenched steel entirely austenitic, then the electrode potential is so modified by the presence of these elements that it is impossible to bring the value obtained into line with the results for common steels. Further discussion of this point may conveniently be deferred till the section devoted to consideration of results.

Cold working or otherwise straining a metal produces a change in the electrode potential of the metal. Fawsitt (17) investigated the electromotive force of cells made up of electrodes of the same metal in the strained and unstrained condition. It appears that with iron the effect of strain of this kind is very small. How far the state of internal strain commonly associated with martensite may affect the electrode potential

potential of martensitic steels is not known. In the present paper, no account will be taken of possible internal strain. It is felt that some confusion has arisen with regard to strain, as if strain by itself were accountable for certain factors particularly noticeable in martensite. Whatever internal strain there may be in martensite, it, like other properties is an effect of the forcible retention of carbon in solution in alpha iron. No doubt, because of this forcible retention, there exists on the part of the carbon, a strong tendency to come out of solution. This tendency may be called strain if need be, but it must not be overlooked that the properties of martensite are directly attributable to the fact that this forced solution has been formed, and that they are in no way attributable to any condition which has arisen because of the formation of this solution. Thus high specific volume, great hardness, high resistance, and what is called internal strain are all equally effects of the forcible retention of large amounts of carbon in alpha iron. That all of these properties vary together and in certain directions is perfectly natural, and cannot be taken as an indication of anything, other than that they are all effects of the same cause. Thus neither high strain nor high specific volume can cause great hardness, but a particular solid solution of carbon in alpha iron formed under certain conditions of cooling can, and does possess all these properties in a marked degree.

This point has perhaps been laboured unduly: but it is felt that, after what has been said with regard to the work of Fawsitt, it is advisable to set out the author's view on this matter. When the electrode potential of martensitic steels comes to be dealt with, everything will be attributed to the solution and precipitation of carbon, it being supposed that internal strain, as an effect of the dissolved carbon, cannot of

of itself give rise to changes in electrode potential. Before leaving this point it may be well to mention that what has just been said does not preclude the possibility of a discontinuous difference in properties between martensite and such a solid solution of carbon in alpha iron, as may be produced under different circumstances. It is hoped that what has been said about the relations of cause and effect as they enter into the consideration of the character of martensite, has been clearly put. Sometimes in dealing with such abstract points, the best efforts end in pure verbalism, though if an analogy from every-day events were permissible, the point could be made briefly and easily.

Potential differences occur at the junction of two liquids. Nernst showed that this arose from the different migration velocities of the ions. Thus, at the junction of two different liquids such as water and hydrochloric acid, the H ions migrate into the water more rapidly than the CL ions, so that the water becomes positively charged with respect to the acid. In the same way a dilute solution of hydrochloric acid in contact with a more concentrated solution will assume a positive charge. Generally, when two solutions of the same nature, but of different concentration, are in contact, the dilute solution will take the charge of the fastest ion.

The nature of the potential difference at liquid junctions can be forseen, and in certain cases it can be calculated. When the liquids in question are solutions of the same salt but of different concentration, or even when they are solutions of different salts of the same metal, the calculation is comparatively easy. But when there is no common ion the calculation can only be accomplished with difficulty.

P.Henderson (18) has devised a formula for calculating the potential at the junction of two solutions containing different anions and cations in different concentrations. He does

does not state, however, which ionic mobilities should be used, nor at what temperature his equation is valid. For these reasons the equation has not been accepted generally, and in this present research the calculation of the potential at the liquid junction has not been attempted.

Although the exact value of the potential at the liquid junction cannot be easily calculated, its direction can be readily found. As was already stated, the most dilute solution will take the sign of the most mobile ion.

In the present investigation the electrode potential was measured against a normal calomel electrode. The iron electrode, which consisted of a steel specimen immersed in ferric chloride, was connected to the calomel by means of a vessel containing normal potassium chloride. The potential difference across the combination was measured by means of a potentiometer.

If the single potential of the calomel electrode and the potential generated at the junction of the potassium chloride and ferric chloride were known, it would be possible to state exactly the electrode potential of the steel in contact with the ferric chloride.

The single potential of the calomel-electrode has been determined by several methods. The value commonly accepted for ordinary temperatures is +.56 volt as found by Lippmann and Helmholtz (19). More recently, Billitzer (20) has obtained a value that differs from the above by as much as .74 volt. In consequence of this uncertainty, Nernst has advised the use of the standard hydrogen electrode, and has suggested that all electrode potential measurements should be referred to this. The International Congress at Berlin (21) recommended that directly measured values should in all cases be given, the auxiliary electrode being a normal calomel cell. This method of stating the results has been followed in the present invest-

investigation, the values given being the electromotive force of the combination. This has been done for more reasons than those just stated. The procedure, by means of which relatively accurate results were ultimately obtained, differs to such an extent from that commonly followed in work of this nature, that it is apparently impossible to reduce the present results to their absolute values.

Therefore the procedure is described in detail and the results set out as they were obtained

SECTION 2. SUB-SECTION B.

ELECTRODE POTENTIAL DETERMINATIONS: EXPERIMENTAL.

EXPERIMENTAL METHOD USED IN DETERMINING THE ELECTRODE POTENTIAL.

As no thorough investigation of the electrode potential of different heat-treated steels has previously been attempted, the arrangement of the experimental method to be used had necessarily to be evolved from first principles. The first principles were only of use in determining the main lines which the investigation should follow. All the details of the method, which led in the end to comparable results, had to be discovered slowly by trial and error.

In the introduction to this section, some of the difficulties which are likely to arise in determining the electrode potential of alloys are described. Considered as an alloy, steel possesses in a peculiar degree all the properties which give rise to difficulties. Many of these difficulties common to all alloys, are, in the case of steel, so pronounced as to be almost insuperable.

Although all alloys contain, in addition to the major components, a number of subsidiary components commonly classed as impurities, the presence of these impurities can usually be neglected. This can be done quite justifiably, because the range of composition, which usually comes within the scope of such work, is sufficiently wide to dwarf the importance of the minor components. With steel, however, the range of composition with respect to the major components is very narrow, and consequently the importance of the numerous subsidiary components becomes of such magnitude as to exercise a noticeable influence on the results. Further, in annealed steels the major components form a mechanical mixture, whereas, at least the greater proportion of the impurities is in solution in the iron. Now the carbide which forms with the ferrite an eutectiferous series, will have little or no effect on the electrode potential. Only that small proportion of the carbon which is actually retained

retained in solution will appreciably influence the results. In annealed steels, therefore, the minor constituents will enter into the electrode potential measurements as factors of considerable importance.

In the series of steels used in the present investigation the content of minor components remains fairly constant throughout each series. So that these components will affect the absolute value of the results to an appreciably greater extent than they will affect the relation existing between the results. This being so, it will be quite correct, after having drawn attention to the possible effect of the impurities on the absolute value of the results, to proceed to discuss the relative meaning of the results in terms of iron and carbon only, bearing in mind, of course, that no comment on the absolute value of the results can be made without reference to these impurities.

The recognition of these things has an important bearing on the experimental method, for they serve to show beyond a doubt, that any effort to discover an appropriate complex electrolyte, in which the relative concentration of the ions is determined by the composition of the alloy, would be sure to end in futility. From this point of view no better electrolyte can be found than a solution of a simple salt of iron. Later other circumstances which entered into the choice of a suitable electrolyte, will be described.

Hatfield determined the electrode potential of numerous alloys including steels, using strong and normal acids as electrolytes. He took readings of the potential at regular intervals after immersion, and gave the initial and final readings. Owing to the rapid chemical action which would take place with such electrolytes, his results do not necessarily represent the true electrode potential of the alloys.

The first specimens used in the present research were similar to those employed in the resistance measurements. These

These long thin specimens were placed in a glass tube so that half an inch of the specimen protruded from each end. By means of sealing wax one end of the tube was sealed and mercury was poured in at the open end to fill the annular space between the specimen and the tube. One end of a copper lead was dipped into this mercury, the other end being connected to a potentiometer. This method of fixing the specimens, although it is the system commonly employed, proved rather tedious and a new method was adopted.

For the new method small cylindrical specimens each $\frac{3}{4}$ inch long by $\frac{1}{4}$ inch diameter were made. One end of the specimen was bored with a $\frac{3}{32}$ inch hole into which a tapered rod fitted. Before each experiment this hole was carefully cleaned by means of fine grade carborundum powder on a rotating rod, the carborundum being finally cleaned from the hole with alcohol and soft wooden pegs. The tapered rod which held the specimen passed through a tight fitting rubber stopper which also served as a stopper for the cell. To the other end of the rod a copper lead was affixed by means of an ordinary terminal. In figure 11 the whole arrangement of specimen, rod, stopper and lead is shown, as well as the arrangement of the half cells and the connecting medium.

From figure 11 it will be seen that the whole cell is made up of two half cells or electrodes, the steel electrode and the calomel electrode, connected by means of a normal solution of potassium chloride. From each electrode a lead goes to the potentiometer, where the potential across the combination is measured.

In order to discover whether a measurable difference existed between steels of different composition which had been treated similarly, and steels of the composition which had been treated differently, the following experiments were made.

Two half-cells, each containing a steel specimen immersed

immersed in a solution of ferric chloride, were connected by means of the potassium chloride solution, and the potential across the combination measured. It was found that a small but measurable potential did exist between steels of different carbon content, and between steels of the same carbon content which had been subjected to dissimilar heat-treatment.

The first attempts to measure the electromotive force of the steel calomel combination were performed with normal ferric chloride as the electrolyte. With the specimen immersed in the electrolyte, readings were taken at definite time intervals. At first the readings fell gradually, but after the first half hour a steady rise commenced and continued until the specimen was withdrawn four hours afterwards. During this time there was no indication of any approach to constancy. Repeated efforts along these lines were made in order to discover whether there was anything characteristic about the fluctuations, but eventually the experiments were discontinued as unlikely to produce comparable results. It was observed that a dark film formed on the specimen during immersion in the electrolyte, and it was apparent that this coating materially affected the results.

To make certain of the effect of this film, a freshly cleaned specimen was partly immersed in the electrolyte and readings taken at definite time intervals. As before, the readings fell gradually. When this gradual fall had continued for some time, the specimen was pushed a little further into the solution so that a clean area was exposed to electrolytic action. This was followed by an immediate rise in the readings. It appeared from this that the first steady fall was due to the formation of the dark film, and the subsequent rise to the slow creep of the liquid up the specimen.

It was thought that perhaps the concentration of the electrolyte was too high and gave rise to an osmotic pressure capable of depositing hydroxide on the specimen. A number of

of more dilute solutions was tried, but the results were substantially the same. Eventually it was recognised that no advantage would accrue from persisting with a method which gave rise to such erratic results.

A dilute solution of ferric nitrate was tried, but as this solution did not prove sufficiently stable and darkened in the cell, the experiments were not proceeded with.

With a dilute solution of ferric sulphate the same general results were obtained. The curve for electromotive force against time showed no definite maximum or minimum, and the same specimen did not give identical results during successive experiments. Using this electrode it was observed that bubbles gathered on the specimen, and under these conditions it was considered fruitless to proceed.

To obviate the formation of bubbles on the specimen a rotating electrode was employed. The movement of the electrode, however, caused the galvanometer spot to waver considerably, and the mean curve obtained by disregarding these fluctuations was the same as before.

Subsequently a ferrous salt, ferrous ammonium sulphate, was experimented with. The readings were much higher than with the ferric salts. A nickel steel with .44 per cent of carbon gave a reading of .6760 to .6832 volt using this ferrous salt, whereas the same specimen gave .4140 to .4146 volt in contact with ferric chloride.

Otherwise the results obtained with the ferrous salt were similar to those obtained hitherto, and as this salt could not be maintained unoxidised, it was considered even more unsuitable than those used previously.

By this time a sufficient number of iron salts had been tried to demonstrate that further experiments along these lines would be unlikely to lead to conclusive ends.

At the end of this phase of the work it was found necessary to consider carefully what had been discovered. The following points were considered to have been established:

1. The electrolyte must be a dilute solution. Concentrated solutions rapidly deposited a film or gave rise to gas bubbles on the electrode.
2. The electrolyte must be a solution of a ferric salt. Ferrous salts, because of their liability to oxidation, proved inconvenient for this particular work, although it is usually against ferrous salts that the electrode potential of iron is measured.
3. The electrolyte must be a solution of a simple salt. The introduction of complex ions might complicate the work.
4. The salt used must be that of a strong acid, as salts like the nitrate were not stable in solution.
5. No final constant reading was to be expected, and there was no character of constancy about the time-potential curves.

From consideration of these points, it was decided to take initial readings only, and to use dilute ferric chloride as an electrolyte. Proceeding on these lines, a number of experiments was performed but the results were far from satisfactory. It was found that the reading was being taken so soon after immersing of the specimen, that the liquid in the cell was still moving. By allowing the liquid to steady itself before taking the reading, greater accuracy was attained. Even then the results varied by as much as .005 volt, which, although not by any means serious, was outside the permissible limits of error in this work.

At this point it became apparent that the condition of the surface of the specimen was of some importance. Polishing on different grades of emery paper altered the value of the

the electrode potential given by any particular specimen. To obviate this difficulty it was necessary to polish every specimen carefully on 000 paper. The sides of the specimens were difficult to polish in such a menner that the degree of polish on bottom and sides would be similar. Experiment showed that even this would affect the results. Consequently it was decided to insulate the sides and polish the bottoms only. Vaseline was used as an insulator. By these means it was found that, by careful standardisation of the procedure, readings which varied by less than .002 volt could be obtained.

The electrode potential of all the quenched and annealed steels was determined by this method and the results plotted. Although the results obtained for the quenched steels were very good, the limits of accuracy were not narrow enough for the annealed steels. There was still room for improvement, and in order that this might be accomplished the whole operation was reconsidered. Each step was carefully examined and so far as could be seen, the most likely source of inaccuracy was the vaseline. Investigation showed that if the coating of vaseline was too thin erratic results were obtained, whereas, if it was too thick some was removed as the specimen passed through the neck of the cell. Further, if the coating did not completely cover the sides they were not completely insulated, while if the vaseline projected over the bottom, air bubbles were liable to be entrapped at the moment of immersion. Recourse was again had to careful polishing of the sides but without better results. Various other methods of insulating the sides were then tried, but all the methods, which included coating with paraffin wax and covering with rubber tubing, proved equally inadequate. The difficulty was finally solved by using shellac varnish. This varnish was carefully applied to the sides of the specimen with a camel hair brush and allowed to dry before immersing in the electrolyte. Throughout the

the remainder of the work this method of insulating was used and gave satisfactory results. All the results obtained with vaseline as an insulator were discarded, and the whole investigation was repeated. The new results did not differ materially from the previous set, but they were much more satisfactory, as the same reading could be obtained regularly.

It was found, however, that the readings obtained for the same specimen were not absolutely invariable, but differed by .001 volt at different times. Further examination of this point revealed the fact that all specimens were affected similarly and in the same degree. That is, a whole series of steels gave results .001 volt higher in the afternoon than in the morning. This was believed to be due to temperature variations in the laboratory, and it was therefore decided to make use of a simple thermostat. The two half cells were therefore immersed in water and a thermometer placed alongside each. The ferric chloride solution used as the electrolyte was also kept at a constant temperature. In this way the effect of temperature was accounted for.

During this time the details of the experimental procedure were gradually being perfected. Again and again the electrode potentials of the annealed and quenched steels were measured, each time with greater accuracy than before. From the time of commencing these experiments to the time when the experimental work was finally perfected, six months had elapsed. It was felt that during this time some change might have taken place in the quenched specimens. Therefore a new set of specimens was prepared and this time the electrode potential was measured immediately after quenching. Of course it required time to determine the electrode potential of a series of steels, and consequently immediately after quenching does not mean the same day. Even during these experiments it was noticed that the readings varied in a progressive manner from day to day. By

By this time it had become apparent that a standard specimen was essential. This specimen of low carbon content and in the annealed state, was measured with each series in order that results obtained from day to day might be compared exactly.

Using this standard it was clear that the newly quenched specimens were altering gradually, so that the electrode potential of the combination became lower.

A most important feature of the experimental procedure was the exact standardisation of every detail. No useful purpose would be served by describing minutely the evolution of the final standardised practice. From each set of measurements made something else was learned, until eventually the final procedure was perfected. This final procedure, by means of which all the electrode potential results included in this thesis were obtained, will now be described.

A large quantity of normal ferric chloride solution was prepared and carefully standardised. From this main solution quantities of decinormal solution were prepared as required in a graduated flask.

The specimens, whose electrode potential was to be determined, were carefully cleaned on the outside with emery paper. The holes in the specimens were cleaned by filling them with a paste of carborundum and pressing the hole on a mild steel rod rotating in a lathe. After the inside of the holes had been satisfactorily cleaned, the specimens were placed in alcohol and the carborundum which adhered to the sides of the holes was removed with soft wooden pegs. A tapered rod, which had been cleaned with fine emery, was now inserted in each hole and driven in firmly. The sides and bottoms of the specimens were then carefully polished on graded emery, special care being taken to have all the bottoms polished equally. After polishing, the specimens were placed on an iron rod and allowed to remain until any heating effect of the polishing had disappeared. A

A thin coating of shellac varnish was applied to the sides with a small camel hair brush. This was done carefully in a strong light so that the sides might be completely covered without allowing any varnish to overlap the end. Any carelessness during this part of the operation was immediately apparent in the results.

The standard specimen was treated in the same way and the whole series placed to dry.

The requisite amount of decinormal ferric chloride having been prepared, the steel half cell was filled. By means of adding hot or cold water to the thermostat, the apparatus was brought to the working temperature of 17° C.

The potentiometer was now set approximately at the reading expected, and the first specimen to be dealt with was fixed to the copper lead. When all was ready the specimen was quickly inserted in the cell in such a way that about half of its length was immersed. Of course, the amount of the specimen immersed depended on the position of the stopper on the rod. This was therefore kept adjusted at the proper place and the length of the specimen immersed was determined without difficulty. It was necessary to make sure at this point that the stopper fitted the cell tightly, otherwise, when the stop cock was opened, a continuous flow of ferric chloride into the potassium chloride would give rise to certain incalculable diffusion potentials.

Whenever the specimen was immersed in the electrolyte and the rubber stopper securely fitted in the cell neck, the stop cock was momentarily opened to allow the pressure inside the cell to eject a drop or two of solution and so establish equilibrium between the internal and external pressure. With the stop cock closed, the outlet tube of the iron half-cell was placed in the potassium chloride. Exactly 10 seconds after the specimen was first immersed in the electrolyte, the stop cock was opened and contact established between the steel and the

the calomel electrodes. After 5 seconds from the opening of the stop-cock, the key was depressed and the approximate reading taken.

It will be remembered that the value of the electrode potential varied continuously from the moment of completing the circuit in the cell. Therefore, if attempts had been made to balance the electromotive force on the potentiometer in the usual manner, the result obtained would have varied according to the time taken to make the balance. Accordingly the method of taking the reading was arranged so that the instantaneous value registered on depressing the key would be obtained. To do this the scale on which the galvanometer spot moved was graduated in terms of potential. All movements of the spot could thus be read in terms of volts. As the potential dropped continuously from the moment of immersion, the potentiometer was set as near as possible below the value expected. Therefore when the key was depressed, the galvanometer spot jumped in the direction of higher potential and then came slowly back. The size of this jump in terms of volts added to the potentiometer reading, gave the instantaneous value of the potential of the combination.

It was not always possible to make an approximate guess at the reading, and sometimes it required many repetitions before the reading for a particular specimen could be determined. Thereafter the reading was repeated again and again to make certain that the correct reading had been determined.

After the key was depressed and the instantaneous reading taken, the stop cock on the iron electrode was turned off and the specimen withdrawn. Without changing the electrolyte another specimen was placed in the cell and the process repeated. After every second specimen the ferric chloride solution was changed. This was done because it had been found that the solution altered during the experiments. Although the electro-

electrolyte was only changed after every second specimen, care was taken that every alternate time a specimen was used it was used with fresh solution.

This procedure was continued until all the specimens in the series had been used once. All the specimens were then cleaned by washing with alcohol to remove the shellac. The whole process of polishing, cleaning the holes, cooling, varnishing and drying was repeated and readings again taken. This second time the potentiometer could be set much nearer to the reading than before.

This process of preparing the specimens and taking the initial readings was repeated until a satisfactory invariable value had been obtained for all the unknown specimens and the standard. Sometimes as many as thirty repetitions were required before the whole series could be passed as determined.

The results obtained by this means are shown in tables 6 to 11 and plotted in figures 12 to 18.

SECTION 2. SUB-SECTION C.

ELECTRODE POTENTIAL DETERMINATIONS: CONSIDERATION OF RESULTS.

ELECTRODE POTENTIAL. CONSIDERATION OF RESULTS.

Before discussing the metallurgical significance of the present results, it is necessary to establish their physical meaning. Without a definite conception of the meaning of the readings, it is impossible to interpret them in metallurgical terms.

Reference to Tables 6 to 10 or Figures 12 to 16, will show that the E.M.F. of the Iron-Ferric Chloride-Calomel combination is always below the E.M.F. commonly assigned to the calomel cell, namely .560 volts. If all other considerations are neglected, these results signify that the E.M.F. at the Iron-Ferric Chloride junction opposes the calomel electrode, so that the E.M.F. of the combination is less than that of the calomel electrode alone. The actual readings are therefore the difference between the E.M.F. at the Iron-Ferric Chloride junction and the E.M.F. of the calomel electrode. Iron, or steel rather, bears the same relation to the calomel electrode as copper, that is the electrode potential of the steels used in the present work, measured under the conditions described in the last section, is negative in sign.

If the results are examined from this point of view, it will be seen that, as the reading for the combination is higher when an annealed steel electrode is used and lower when a quenched steel electrode is used, the negative potential, at the junction between a quenched steel and ferric chloride, is greater that the negative potential at the junction between an annealed steel and ferric chloride. Further, it will be seen that the negative potential of quenched steels increases with the carbon content.

The purpose of this examination of the meaning of the results is to discover in what way the activity of a steel electrode is affected by dissolved carbon. The positive

positive potential of an electrode is a measure of its activity; the negative potential a measure of its passiveness. So a high negative potential means low activity, and a low negative potential means high activity. The direct readings of the E.M.F. of the Iron-Ferric Chloride-Calomel combination are therefore proportional to the activity of the electrode. Annealed steels are therefore more active than quenched steels, and quenched low carbon steels are more active than quenched high carbon steels. The activity of the steel electrode is reduced by dissolved carbon.

It was pointed out in the introduction to Electrode Potential Measurements (Sub-section A) that the potential of a composite electrode is determined entirely by the most active constituent, independent of the relative amounts of the various constituents. If, therefore, a steel electrode contains areas of varying activity, the potential observed will be that of the most active area.

Assuming that there is more carbon in solution in a quenched steel than in an annealed steel, it has been shown that carbon in solution reduces the activity of steel.

Therefore, if in the same steel specimen there are areas containing different amounts of carbon in solution, that area in which the carbon content is least, will determine the electrode potential of the steel. If the foregoing is true, then in all cases the electrode potential measured is the potential of that area in which the carbon concentration is lowest. This area will subsequently be referred to as the least concentrated area.

To verify the above arguments, a number of experiments were performed with two half cells each containing a steel electrode in contact with decinormal ferric chloride. Two steels of widely different carbon content quenched from the same temperature were used. It was found that the current

current flowed in the external circuit from the high carbon steel to the low carbon steel, and therefore in the internal circuit from the low carbon steel to the high carbon steel. This was repeated many times using different specimens, and always with the same result. The low carbon steels are therefore more active than the high carbon steels, and the annealed steels more active than the quenched.

Before leaving this point it may be mentioned that Bronweiski ( ) found that in a cell made up of a quenched steel and a pure iron electrode, the current flowed in the external circuit from the steel to the iron. This is in agreement with the above experiments.

From consideration of the electrode potential results as a whole, it is evident that the potential measured is that of iron modified by carbon, and probably some impurities, in solution. As no extensive investigations of the electrode potential of quenched steels has hitherto been carried out, there is no way of approaching these results other than that shown by the results themselves. Thus so far, it has been shown that the readings are the electrode potential of iron modified by dissolved carbon, and that the area, in which concentration of carbon is lowest, determines the reading.

Another point which requires examination is whether the appearance of gamma iron in certain quenched steel affects the readings. This point has already been referred to in the introduction, but a more specific treatment is now required. It is unlikely that gamma iron is retained in the lowest carbon steels, so it is reasonable to assume that the electrode potential of these steels is not that of gamma iron. If gamma iron is less active than alpha iron, its appearance will not affect the electrode potential readings unless the steel is entirely austenitic when quenched. If it is more active than alpha iron, its appearance should be accompanied by an

an upward break in the curve of electrode potential against carbon percentage. Reference to Figures 12 to 16 will show that the only break in these curves is downwards, and therefore cannot be due to the appearance of a more active constituent at a certain content. Gamma iron, as austenite, does not appear to influence the results.

Discussion of the metallurgical significance of the electrode potential results shown in tables 6 to 10 and in Figures 12 to 18. may now be attempted. Two factors serve to make difficult the interpretation of these results; the dependence of the results on the carbon content of the least concentrated area; and the logarithmic relation between electrode potential and carbon in solution. These factors make the electrode potential results so different from the results obtained by the more common methods of investigation, that deductions must be very carefully drawn.

The tempering results shown in Figures 14 to 16 and in Tables 6 to 9 are comparatively straightforward. They show that the amount of carbon in solution in the ferrite decreases gradually as the tempering temperature is raised. As tempering was continued for 3 hours at each temperature and for 6 hours at 240° C., it is apparent that time and temperature are not interchangeable factors. At each tempering temperature the steel appears to attain a certain state from which it departs very slowly, if at all.

Similar indications were obtained during the measurement of the specific volume and of the electrical resistance of tempered steels, and are discussed in section 1, sub-section B, and in section 4, sub-section B. These results, however, do not point so clearly to the nature of this characteristic state. They might quite reasonably be attributed to the effect of state of division.

The electrode potential results do not admit of an

an explanation based on state of division alone. They show that at each temperature the amount of carbon in solution in the tempering steel reaches a certain limiting value.

It is well known that tempering at different temperatures produces a structure of varying degrees of fineness, and it is commonly accepted that this variation in the fineness of the structure is responsible for differences in physical properties. From the electrode potential results it appears that fineness of structure is not an entirely independent factor, but is associated with the amount of carbon in solution.

From analogy with the foregoing it seems that the different degrees of structural fineness, obtained by cooling a steel at different rates through the change point, may also be associated with variations in the amount of carbon in solution. In fact, unless two distinct varieties of troostite and sorbite are to be recognised, it is essential that the similarity between tempering at different temperatures and cooling at different rates be admitted.

This conception introduces an element of continuity into heat-treatment. It suggests a continuous variation in the amount of dissolved carbon accompanying the gradation of structural fineness from granular pearlite to troostite, so that tempering at different temperatures or cooling at different rates from high temperatures, permits the constituents to separate from each other with varying degrees of completeness. And it suggests that completeness of separation is associated with fineness of division, in such a manner as to make it impossible to ascertain whether certain properties are effects of dissolved carbon or of fineness of division. Neither of these factors is the cause of the other: they are concomitant. Neither of them is the cause of the different properties of pearlite, sorbite, troosto-sorbite or troostite steels: they are equally responsible.

Recognition of the similarity between the various constituents produced under different conditions of cooling, and those produced under different conditions of reheating, does not constitute a disagreement with the views of Whiteley (23) and Hallimond (14). The former recognises microscopically two different troostites, distinguishable from each other by their mode of occurrence. The latter recognises two troostites distinguishable by the manner of their origin. The present view is that the ultimate constitution and properties are identical whatever difference there may be in mode of origin or occurrence.

The curves shown in Figures 14A and 16 give the impression that the carbon comes out of solution in approximately equal increments, as the temperature of tempering is raised by equal steps. This is exactly how the results work out. When the electrode potentials determined after tempering at different temperatures are plotted against tempering temperature, it is clear that for each steel a linear relation exists between electrode potential and tempering temperature. The results were plotted on these axis to emphasise this relationship, but as it is sufficiently evident from Figures 14A and 16 these additional figures are not included.

From Figures 19 and 20, in which the electrical resistance of steels quenched and tempered at 240° C. is shown, it appears that tempering has proceeded a fair way to completion during a few hours at 240° C. This apparent disagreement between the electrical resistance and electrode potential measurements, has its origin in the different relations which exist between these properties on the one hand, and composition on the other. Within the limits of composition of the steels employed, electrical resistance is a linear function of the carbon in solution, whereas electrode potential is evidently a logarithmic function of carbon in solution. The first small amounts

amounts of carbon dissolved have a notable effect on the electrode potential. As the total carbon in solution becomes considerable, further increases do not affect the electrode potential to the same extent. The converse is also true. Thus, when the carbon content is considerable, the separation of an appreciable amount will have the same effect on electrode potential as the separation of a minute quantity from a solution containing very little.

The electrode potential curves shown do not give a very clear idea of this logarithmic relation, but other considerations may be taken into account. During the determination of the electrode potential of annealed steels, some difficulty was experienced in finding definite readings capable of repetition. Each time the specimens were repolished the electrode potential varied a little. It did not vary continuously in any direction but oscillated about a mean. The steel with the lowest carbon content (.22% Carbon) behaved in a similar manner when quenched, and different specimens of this steel quenched under identical conditions, gave rise to different electrode potentials. The remainder of the steels did not behave in this manner when quenched, nor was there any appreciable difference in the electrode potential of different specimens of the same steel quenched in the same way.

It appeared that each time a specimen was repolished an area of slightly different carbon content was exposed to the electrolyte. The difference in the effect of this procedure on annealed and on quenched steels, can only be attributed to one or other of the two following causes.

(1) Either quenched steels are more homogeneous than annealed.

(2) Or the effect of small variations in the amount of carbon in solution is more when the total amount of carbon in solution is small, and less when the total

total amount of carbon is large.

Taking this into consideration with what has already been said about the effect of tempering on electrode potential, it is obvious that the second of these explanations is the more probable.

Returning to the curves showing the effect of tempering on the electrode potential of quenched steels, it is evident that although the electrode potential is a linear function of the tempering temperature, the carbon which separates from solution is not similarly related, but varies as the logarithm of the tempering temperature.

The curves, as they are plotted, give therefore a disproportionate idea of the amount of carbon which remains in solution after tempering at different temperatures: but bearing the foregoing in mind it is still possible to appreciate their qualitative significance.

At the beginning of this discussion it was pointed out that if the steel under examination was not homogeneous, the electrode potential obtained would be that of the area in which the amount of dissolved carbon was least. The importance of this in discussing these results is obvious. But unfortunately the results for the quenched steels do not show whether the carbon is uniformly distributed throughout the steel or not. If the carbon is distributed uniformly in the quenched steels and remains thus distributed throughout the tempering experiments, it is possible to discover, by drawing horizontals in Figures 14A and 16, how much carbon is in solution in any particular steel after tempering at any particular temperature. For example, on this assumption the carbon steel containing 1.73 per cent of carbon, after tempering at 240° C., appears to contain .7 per cent of carbon in solution. After tempering at 360° C. for 3 hours, it appears to contain .48 per cent of carbon in solution, and after tempering at 450° C. about .3 per

per cent of carbon. These figures are obtained by finding the point at which a horizontal line from the electrode potential of this steel after tempering, meets the line representing the electrode potential of quenched steels. By producing upwards the line representing the electrode potential of quenched steels, it is possible to determine the amount of carbon in solution in any steel after tempering at any temperature.

But if the steels as quenched and tempered are not homogeneous, that is, if the carbon is not uniformly distributed throughout the alpha iron, this method of giving quantitative meaning to the results is worthless. For if there are areas of different carbon content in the steels, and if the area of least carbon content determines the electrode potential, all that may be said is that the least concentrated area in the steel containing 1.73 per cent of carbon and tempered at 240° C., contains the same amount of carbon in solution as is contained in the least concentrated area of the steel with .7 per cent of carbon in the quenched state. As the amount of carbon dissolved in the least concentrated area of the quenched steel containing .7 per cent of carbon is not known and cannot readily be discovered, no quantitative deductions can be made.

It is evident that the distribution of carbon in quenched and tempered steels has a very intimate connection with the results under consideration. Therefore, as the results themselves do not indicate the nature of the carbon distribution, some other considerations may be taken into account.

Whiteley (23) states that the martensite in quenched steels is perfectly homogeneous, that is, the carbon is distributed uniformly throughout the alpha iron. He bases this statement on results obtained by tempering martensitic steels at 400° C., when they appeared uniformly black under the microscope. In drawing this conclusion from microscopic observations made on troostite produced by tempering martensite, Whiteley

Whiteley is assuming that uniform blackness means uniform distribution of carbon in the troostite. But it has been observed that steels of different carbon content become almost equally black when tempered to produce troostite. Thus, uniform darkness is not irrefutable evidence of uniform distribution of carbon.

Rawdon and Epstein (25) and Hannemann state that the distribution of carbon in martensite is not uniform. In these two investigations the experimental method was substantially the same. The unequal distribution of the carbon was observed by tempering martensitic steels at 650° C., so that the carbide would be obtained in microscopically distinguishable accumulations. When the sorbitic structure produced by this treatment was examined, areas of different carbon content were recognised.

Rawdon and Epstein and Hannemann conclude that the unequal distribution of carbide observed in these tempered steels results from unequal distribution of the carbon in the original martensite.

Similar experiments were performed by Whiteley, who obtained identical results. His micro-photographs show clearly that areas of comparatively free ferrite and areas rich in cementite exist in martensitic steels tempered at 650° C. He arrives at a different conclusion from the other investigators, and states that " there seems to be no alternative to the conclusion that these ferrite areas which appear in tempering are due to an exceedingly rapid migration of the carbide particles formed in the earlier stages." To prove this Whiteley endeavoured to obtain micro-photographs of intermediate stages in the migration. He found that with a total tempering time of only 30 seconds the migration was complete, and it was therefore impossible to obtain evidence of intermediate stages in the process. The migration, if it does take place at the tempering temperature, must do so instantaneously.

All these investigators obtained the same results but they arrived at different conclusions. The results obtained by microscopic examination of tempered martensitic steels can be explained by either assumption, that is, heterogeneity in the tempered specimen may, with equal justification, be attributed to heterogeneity in the quenched specimen or to rapid migration of the carbon during tempering. This, of course, holds true of all experiments performed on tempered specimens. It is never clear from the results obtained what proportion of the observed changes are attributable to something intrinsic in the quenched steel, and what proportion to the tempering process. The same difficulty arises with quenched steels: it is frequently impossible to decide whether the observed phenomena are the result of some condition existing at the quenching temperature, or of some change during cooling.

There are objections to both views on the origin of the heterogeneity found in steels tempered at 650° C.

Whiteley has stated the case against the view that heterogeneity exists in the quenched specimen. He says that the idea involves diffusion of carbon at low temperature during the formation of martensite,(of course, the heterogeneity may exist in the steel at the quenching temperature, but this is unlikely); that areas of free ferrite were also noted in tempered hypereutectoid steels; and that when martensite was tempered to produce troostite, a uniformly black structure resulted.

He was, however, unable to suggest any sufficient reason why the carbide should migrate so rapidly at 650° C. The suggestion that two varieties of alpha iron result from the breakdown of martensite is hardly an adequate explanation.

The case against the view that the carbon is uniformly distributed in quenched steels may be stated as follows:-

If the carbon is distributed evenly throughout the steel

steel so that the carbon content of martensite is absolutely uniform, is the carbon content of austenite (where this constituent is retained) the same as that of martensite? If the carbon content of austenite is the same as that of martensite, how is austenite retained? The pressure arising from the expansion that accompanies the formation of martensite, does not account for the retention of austenite. For if austenite were retained by pressure alone, then during tempering, austenite would decompose as quickly as the pressure was removed, that is, it would temper at the same rate as martensite. In section 4, sub-sections C and D, it is shown that austenite tempers much more slowly than martensite. Therefore pressure alone does not cause the retention of austenite.

Pressure, of course, does play a part in preventing the breakdown of austenite during cooling, and its effect is most evident in the results obtained by quenching specimens of such dimensions that the outside is cooled quicker than the inside. When such a specimen of hypoeutectoid steel is quenched, it is commonly found that, whereas the outside of the specimen is martensite, the inside is troosto-martensite or troostite. When a hypereutectoid steel is used for this experiment, it is frequently martensitic on the outside and austenitic in the inside. With the hypoeutectoid specimen, the arrangement of the constituents along the temperature gradient is in accordance with the well known effect of the rate of cooling on the constituents produced by quenching. With the hypereutectoid specimen the arrangement is the reverse of what might be expected from the effect of rate of cooling only. The expansion which accompanies the change from austenite to martensite on the outside, apparently prevents the change in the inside. Pressure is therefore a factor in the retention of austenite, but as was shown earlier, it is not alone responsible.

There must therefore be some difference in the carbon

carbon content of martensite and austenite - as Hanneman tried to show. And if martensite is homogeneous, yet contains less carbon than austenite, it follows that the steel has become divided into two distinct solid solutions, the carbon contents of which are different. The line of demarcation between these two solid solutions must be definite, and whether this separation into two solid solutions is assumed to take place during cooling, or to exist in the steel before quenching, it is nevertheless difficult to explain.

There are therefore objections to both views on the origin of the heterogeneity produced by tempering at 650° C., and although one of them must be correct, it is difficult to decide which. The results obtained in the present research do not assist in deciding this point, but the writer is inclined to favour the view that carbon is unevenly distributed in quenched steels. Therefore, as there is no evidence to show what relation the carbon content of the least concentrated area bears to the carbon content of the steel, the electrode potential results obtained for quenched and tempered steels cannot be interpretated quantitatively.

To ascertain the effect of the time of tempering, the steels were tempered at 240° C. for two periods of 3 hours. From Figures 14A and 16, where the effect of this treatment is shown, it appears that electrode potential is but little affected by prolonging tempering.

Raising the tempering temperature increased the electrode potential. Tempering at 680° C., however, produced an appreciable decrease in the electrode potential of carbon steels, and this effect was enhanced when the steels were quenched from the tempering temperature instead of being cooled in the furnace. Tempering at 730° C. had a similar effect on the electrode potential of chromium steels. With the latter steels the effect of this treatment was most marked when the specimens

specimens were quenched after tempering.

Campbell, when measuring the specific resistance of tempered steels, noticed that, whereas the resistance decreased progressively as the tempering temperature was raised to 650° C., tempering at 650° C. caused a relative increase.(27)

This decrease in the electrode potential produced by tempering, indicates that carbon is going into solution in the alpha iron within some temperature range just below the change point. Further discussion of this point may be reserved until after the results obtained for annealed steels have been described.

In Figures 14 and 15 the electrode potentials of carbon and chromium steels quenched from 1000° C. and tempered, are shown. The electrode potentials of these steels after tempering at 240° C. approached so closely to the electrode potentials of the same steels quenched from 1100° C and tempered at 240° C., that it was not considered worth while to proceed with them. These results indicate that after tempering at 240° C., the steels attain the same state whether they were originally quenched from 1000° C. or 1100° C.

In Figures 12 and 13 the electrode potentials of carbon and chromium steels quenched from 1000° C. and 1100° C. are compared. The peculiar shape of these curves may be mentioned. A definite break, the exact significance of which is not apparent, occurs in the neighbourhood of the eutectoid composition.

It has already been explained why this break cannot be attributed to the appearance of austenite, and it now remains to discover, if possible, to what it is due. The break persists in most of the curves which show electrode potential of the steels after tempering. Consequently, it cannot be due to any particular arrangement of the constituents in the quenched steel of this composition, nor to strain. It cannot be due to any compound of iron and carbon in the atomic pro-

proportions represented by this composition, for, after tempering under conditions that produce a microscopically distinguishable separation of iron and carbon, the break still persists. It is probably due to some peculiar relationship between the carbon in solution and the carbon in the steel.

These measurements of the electrode potential of quenched steels were repeated many times, and four different sets of specimens were used. All the results were substantially the same, and the break was always observed at the same point.

There is little difference between the electrode potential of carbon and chromium steels as quenched. As all the chromium is in solution when the steels are quenched from 1000° C. or 1100° C., it is evident that chromium has either very little effect on electrode potential or is concentrated in areas which do not give rise to the readings obtained.

The nickel steels examined gave very different electrode potential readings from the carbon and chromium steels, and there is no apparent reason why chromium should not lower the electrode potential as nickel and carbon do.

The electrode potential of annealed chromium steels is below that of annealed carbon steels. This indicates that there is some chromium in solution in the ferrite of annealed chromium steels, or that more carbon is retained in solution because of the chromium. As it is generally believed that there is more chromium in solution in quenched steels than in annealed - and other results obtained during this investigation confirm this belief - it is surprising that there is less difference between the electrode potential of carbon and chromium steels in the quenched state than in the annealed state.

From analogy with the electrode potential of annealed and quenched nickel steels and annealed chromium steels, and from analogy with the electrical resistance results (Section 3) and specific volume results (Section 2), the chromium steels

steels when quenched should give appreciably different electrode potentials from the carbon steels.

The electrode potential of quenched chromium steels is not greatly different from that of the carbon steels, and when all relevant facts are taken into account, it is fairly evident that these results arise from an unequal distribution of the chromium. If the chromium is not uniformly distributed in the quenched steels, but is concentrated in certain areas, thus leaving other areas comparatively free from chromium, then, as chromium in solution lowers the activity of the iron, the electrode potential measurement will be that of the areas in which no chromium is retained.

If the ferrite in annealed carbon steels were pure alpha iron or alpha iron containing a fixed amount of impurities in solution, the electrode potentials of all steels in the series would be similar; and the curve representing the variation of electrode potential with carbon content would be a straight line parallel to the composition axis. In Figure 17 the electrode potential of annealed carbon steels is shown. The curve representing the electrode potential after heating for 1 hour at 850° C. will be considered first. This is the treatment which was generally given to annealed specimens. The curves that represent the electrode potential of steels annealed in various other ways, will be considered as special cases.

From this curve it will be seen that the electrode potential becomes less as the carbon content increases. This means that the ferrite areas which give rise to the electrode potential are not of constant composition throughout the series of steels, but vary in composition with the composition of the steel. The nature of this variation indicates that the ferrite areas, which give rise to the electrode potential, become less pure as the carbon content of the steel is increased. There is every reason to believe that this variation is due to an

an increase in the carbon content of the ferrite areas.

There is a convenient explanation of this phenomenon. If ferrite forms from austenite by a process of selective freezing similar to that by which heterogeneous crystals are formed during the freezing of a solid solution, the first areas of ferrite to form will contain relatively less carbon than those formed subsequently; and further, the areas of ferrite first formed will be purer the lower the carbon content of the steel.

If this conception of the process of formation of ferrite is correct, the shape of the electrode potential curve is explained. For, if the ferrite is heterogeneous, and if concentration of carbon in the least concentrated area increases as the carbon content of the steel is raised, the electrode potential of annealed steels will become lower as the carbon content becomes greater.

It is, of course, commonly admitted that the selective formation of ferrite is a thermodynamical necessity; but it is usually held that the effect of this is so infinitesimal as to be negligible in practice. It appears, however, that the electrode potential measurements are susceptible to these small variations in the carbon content of ferrite. This susceptibility is not due entirely to the sensitivity of the method, but is partly accounted for by the selective nature of electrode potential measurements, which enable the most active area to be isolated from the less active areas surrounding it. No other method of research at present made use of, except microscopical examination, is capable of thus selecting certain areas for investigation. But the microscope, though usually suitable, is incapable of detecting variations in the carbon content of ferrite.

This very selective action of electrode potential

potential measurement makes the results obtained by it unsuitable for quantitative interpretation. For, as was pointed out previously, if the electrode potential observed is that of a certain area in a heterogeneous specimen, and if the concentration gradient throughout the specimen is unknown, the coordination of carbon in solution and electrode potential is impossible. Therefore the line in the iron-carbon diagram, representing the composition of the ferrite which is in equilibrium with austenite of a given composition, cannot be drawn. But from considerations based on other aspects of the work, it may be deduced that this line lies very close to the temperature axis.

This idea of the selective formation of ferrite from austenite is sufficient to explain the variation in electrode potential up to .9 per cent of carbon. Above .9 per cent of carbon there is no free ferrite, and the ferrite which is a constituent of pearlite, should be of identical composition in all steels. The electrode potential of all steels containing pearlite, and pearlite with cementite, should be the same. As the line representing the electrode potential of carbon steels annealed for 1 hour at 850° C. will show, the electrode potential-composition curve is practically parallel to the composition axis above .9 per cent of carbon. What little variation there is might be attributed to the effect of the rate of cooling on the separation of free cementite from austenite and on the simultaneous separation of cementite and ferrite at the eutectoid. Thus, with the same rate of cooling, the temperature at which pearlite forms may be a little lower when the amount of free cementite that must separate before the simultaneous separation can commence, is greater. Formation of pearlite at a slightly lower temperature would be accompanied by a less complete separation of the constituents, and the ferrite so formed would contain more carbon.

The curve representing the electrode potential of chromium steels annealed for 1 hour at 850° C. is shown in Figure 18. It is similar in form to the curve for carbon steels and is undoubtedly due to the same causes.

The curves representing the electrode potential of carbon and chromium steels annealed at 850° C. for 1 hour are approximately parallel, and the constant difference between them is apparently due to a constant amount of chromium in the ferrite of the chromium steels.

A comparison may be made between the electrode potential of the steels annealed in this way, and that of the steels fully tempered after quenching from 1100° C. Below 1.2 per cent of carbon, the electrode potentials of the carbon steels fully tempered are above those of the same steels annealed. Above 1.2 per cent of carbon the annealed steels give the highest electrode potential. The curves cross at the point where austenite first appears in the quenched steels. After tempering at 240° C., 350° C., 450° C., 550° C., and finally at 650° C., the steels must have reached a condition very far removed from the quenched state, and therefore the relation between the constitution of the quenched steels and the electrode potential of the tempered steels, is not immediately apparent. But there are many indications, for example the persistence of the break, that the condition produced by quenching influences the result of tempering, although all that is typical of the quenched state has disappeared. Not only the persistence of the break, but the general shape of the curves obtained after tempering, lends support to this idea.

If the state attained after tempering were not affected by the condition which existed before tempering commenced, the shape of the electrode-potential composition curve would be different. For after tempering for 6 hours at 240° C. and

and 3 hours at 350° C., every carbon steel should consist of carbide and ferrite in a fine state of division. There is no reason why the carbon in solution in the ferrite should vary with the carbon content of the steel, that is, no reason other than that some residual effect of the quenched state still persists.

How the conditions, which existed in the quenched steels, affect the tempered steels cannot be easily explained. The electrical resistance and specific volume results give no indication of this phenomenon, but neither of these methods is capable of selecting certain areas; they are influenced only by the total condition existing. Thus, whatever the cause of the electrode potential results obtained, it seems to be in some way connected with the selective action of this method.

As already stated, the electrode potential of the steels with less than 1.2 per cent of carbon is higher than that of the annealed steels. This indicates that in those steels, prolonged tempering at 650° C. produces purer ferrite than annealing for one hour at 850° C. The effect of tempering at 680° C. has already been referred to, and was ascribed to increased solubility of carbon in alpha iron at temperatures near the change point. This increased solubility is not clearly revealed by measuring the electrode potential of tempered steels, for it is not until the carbon retained in solution by quenching has completely separated, that the effect of temperature on the solubility of carbon in alpha iron appears. So far as these results are evidence of this increased solubility, they indicate the existence of a line in the iron-carbon diagram representing variation with temperature of the carbon content of ferrite.

In Figures 17 and 18 there are shown, in addition to the curves for steels annealed for one hour at 850° C., a number of curves representing the electrode potential of carbon and

and chromium steels skowly cooled after various heat treatments. Most of these treatments were carried out to obtain electrode potential results corresponding to some specific volume determinations made by Mr. M.S. Fisher. Heat treatment, similar to that given to the electrode potential specimens, produced graphitisation in the high carbon specific volume specimens. It was hoped that the electrode potential determinations would assist in explaining the process of graphitisation. It is now clear that the electrode potential results have no connection with graphitisation.

The graphitisation obtained by heating high carbon steels to 1100° C., cooling slowly to 850° C. and soaking there, does not in any way affect the composition of the ferrite which is part of the eutectoid. As it is the eutectoid ferrite alone that determines the electrode potential of annealed hypereutectoid steels, it is evident that the results are uninfluenced whether the excess carbon appears as graphite or as cementite. No graphite was produced in hypoeutectoid steels by this treatment, and the specific volume results were not affected. Therefore the electrode potential of hypoeutectoid steels should not be modified by graphitisation.

Although determinations of the electrode potential of steels heat treated to produce graphite, are of no assistance in explaining the process of graphitisation, they are still sufficiently anomalous to be worthy of note.

In Figure 17 the curves representing the electrode potential of carbon steels heated to 1100° C., cooled slowly to 850° C. and soaked there for 3 and 6 hours respectively are shown. These two curves are sufficiently alike to be considered together. Below .65 per cent of carbon. the electrode potential curve of steels heat treated in this way lies above the curve representing the steels soaked for 1 hour at 850° C. This indicates that heating to 1100° C. followed by slow cooling

cooling and soaking at 850° C., produces purer ferrite. According to the explanation already given of the selective formation of ferrite, this purer ferrite could only be produced if the conditions of cooling permitted a nearer approach to equilibrium. Areas of purer ferrite could be produced either by slower cooling through the temperature at which the first ferrite areas form, or by soaking in the neighbourhood of that temperature. Evidence of this purer ferrite was only obtained in two steels (.22 and .48 per cent of carbon), so that either of the above suggestions might meet the facts.

Above .9 per cent of carbon, the curves representing the electrode potential of the steels heat treated in the manner described, lie below the curve for the annealed steels. This shows that in the former steels the carbon content of the eutectoid ferrite is greater than in the latter. This would be accounted for by a difference in the rate of cooling through the eutectoid temperature.

Unfortunately the importance of the rate of cooling at various stages in the cooling process was not realised at the time these experiments were performed, so that, although it may be suggested that the observed results are caused by variation in the cooling rate, they cannot be definitely attributed to it. It is now apparent that to obtain strictly comparable results for the electrode potential of steels heat treated differently and slowly cooled, a complete record of the cooling must be kept.

The results obtained for the carbon steels in the rolled condition are further evidence of the importance of the rate of cooling.

Ordinarily all of these slow cooling treatments would be designated annealing, and these electrode potential determinations show clearly what a multiplicity of conditions this term may include.

Apart from graphitisation, which microscopic examination or specific volume determinations would immediately reveal, there is a number of more obscure differences in internal constitution, which may be produced by soaking at different temperatures and cooling at different rates. It is impossible to say how far these effects would influence methods of investigation less sensitive than electrode potential determinations.

What has been said about the electrode potential of slowly cooled carbon steels is equally applicable to the chromium steels (Figure 18). Some of the results obtained by soaking chromium steels for various times at 850° C. are very peculiar, but without a proper record of the cooling process they cannot be adequately explained.

In conclusion it may be said that although the electrode potential determinations have not so far revealed anything of pronounced importance, they indicate that many useful data might be obtained by this method. Although the method used in making these measurements does not take account of many theoretical requirements, it is nevertheless capable of determining a characteristic value, capable of repetition, for each particular steel in each particular condition.

The results obtained are so different from those obtained by the methods of research commonly used in metallurgy, that they were somewhat bewildering at first. When the method had been standardised so that measurements could be repeated with remarkable accuracy, a great number of determinations had then to be made on different steels in different conditions, before the meaning of the results was even obscurely realised. Practically all the results included in this thesis were obtained before the relations between heat treatment, composition and electrode potential could be properly worked out. Therefore what the method is capable of has not yet been adequately proved. But with a knowledge of its limitations, its sensitivity

sensitivity and its meaning, the method of measuring the electrode potential should prove a useful aid to metallurgists, especially in elucidating the phenomena of quenching and tempering.

SECTION 3. SUB-SECTION A.

ELECTRICAL RESISTANCE DETERMINATIONS: INTRODUCTION.

ELECTRICAL RESISTANCE OF CARBON AND ALLOY STEELS.

In this section a method of determining the electrical resistance of steel is described. The method is substantially the same as that familiarly employed for this purpose and is based on the common principles. Some of the details, however, are original and designed to ensure accuracy.

By means of this method the resistance of a number of steels as quenched from different temperatures and as annealed is determined, and the results so obtained are discussed.

PRELIMINARY REMARKS.

The electrical resistance of metals and alloys has long been regarded as an important physical property. Many investigations have been carried out by metallurgists and physicists with varying objects. Sometimes these enquiries have been pursued with no other end than to discover the resistance of a particular material, in order that its suitability for electrical purposes may be determined. More frequently, however, these resistance measurements have been designed to disclose the internal arrangement of the constituents in the material, and the changes produced in them by certain treatment.

The first exhaustive investigation in this branch of physical metallurgy was carried out by Matthiessen (28) who, in the course of many years' work, collected sufficient data to establish the fundamental laws connecting resistance and constitution. Matthiessen did not formulate these laws at the time. It was left to Le Chatelier (29) to draw the main conclusions from the results of Matthiessen and from data accumulated by himself. Le Chatelier confirmed two distinct groups of alloys in which the resistance varied according to certain laws.

In the first group are included all alloys in which the components are mutually insoluble in the solid state. The

The electrical conductivity is here a linear function of the volume concentration of the two components. Some modification of this law was suggested by Guertler (30) and Kurnakow and Zemezuzny (31). These investigators pointed out that the distribution of the insoluble constituents would, under certain circumstances, restrict the application of the law. Subsequent investigations have substantiated these views, and it is now generally admitted that, when one of the constituents is a poor conductor and is distributed mainly on the grain boundaries, its effect on the conductivity is notably different from what it would be if the two metals formed an eutectiferous series.

In the second group are those alloys whose components form a continuous series of solid solutions. The conductivity curve in this case is U shaped.

The third group where the metals form compounds was first stated by Guertler. He affirms that the formation of a compound is accompanied by a distinct discontinuity in the conductivity-concentration curve, and that the compounds have less conductivity than the most conductive component. From this it is evident that good conductors are unlikely to be produced by mixing indifferent or bad conductors.

The effect of foreign elements on the resistance of iron has been sutdied by numerous investigators. Barrett, Brown and Hadfield (32) carried out a comprehensive investigation of the resistance of 110 binary, ternary and quaternary steels. From consideration of their results, they suggested that the effect of another element on the resistance of pure iron is inversely proportional to its atomic weight. They further concluded that the effect of a number of elements was additive, each element producing its characteristic effect, whether present alone or in conjunction with other elements.

Benedicks (33) improved on the speculations of the above

above authors. He formulated a complete explanation of the quantitative effect of other elements on the resistance of iron. This law of Benedicks states that equivalent quantities of foreign elements dissolved in iron produce equal changes in the resistance. By extrapolating the line representing the effect of carbon on the specific resistance of quenched steels to meet the resistance axis at 0 per cent carbon, it was found that the specific resistance or iron was 7.6 microhms. Further, from the slope of this same line it was found that 1 per cent of carbon increased the specific resistance by 26.8 microhms. Benedicks therefore summarised the effect of the various elements in the expression

$$R = 7.6 + 26.8 \Sigma C.$$

where R = specific resistance of any steel.

and C. = carbon equivalent of all elements, other than iron, in the steel.

Since this theory was first advanced numerous investigators have in turn endeavoured to confirm or refute it. The specific reisstance of 7.6 microhms conferred on pure iron has been specially criticised. All attempts to measure the resistance of pure iron have led to much higher values. There is, however, one notable exception. Thomson obtained 7.56 microhms using a rolled bar of pure crucible iron, and reduced this value to 7.15 by drastic annealing.

Benedicks himself was prepared to admit that the value obtained by extrapolation was somewhat lower than any value likely to be obtained directly. He therefore suggested that, according to his theory, .004 per cent of hydrogen dissolved in the pure iron would raise the resistance from 7.6 to 9.5 microhms, the value usually obtained. But Sieverts (34) heated iron wire in hydrogen and obtained no change in resistance.

It is an interesting fact that the value 9.5 microhms for

for the specific resistance of iron was first obtained by Le Chatelier by extrapolation from a line representing the specific resistance of annealed steels, and the value 7.6 microhms by Benedicks by extrapolation of a line representing the specific resistance of quenched steels. From the electrical resistance results obtained during the present investigation, both of these values are approximately confirmed. The values for the specific resistance of carbon steels when plotted, lie on two lines representing the quenched and annealed states respectively. These lines cross about .1 per cent of carbon and meet the axis in the neighbourhood of the above values. A similar feature characterises the specific volume curves. Extrapolation is therefore a doubtful expedient in this respect. Later, the views which the present results engender, will be explained.

The present research on resistance was executed in the first place as a preliminary to the continuous measurement of the change of resistance during tempering. Besides the primary object of determining the specific resistance of the steels examined, the work was designed to discover with what degree of accuracy the resistance could be measured, to what extent small variations in temperature affected the results, whether or not slight heterogeneity in the specimens produced measurable alteration in the readings, and within what limits of accuracy the cross-section of specimens could be measured.

METHOD OF MEASUREMENT EMPLOYED.

The resistivity determinations were carried out by passing a known current through a long narrow cylindrical specimen, and measuring the voltage drop between the knife edges, which were a known distance apart and in contact with the specimen. The specific resistance, at the temperature of the experiment, was obtained by calculation from the diameter of the specimen and the distance between the knife edges.

SECTION 3. SUB-SECTION B.

ELECTRICAL RESISTANCE DETERMINATIONS: EXPERIMENTAL.

SPECIMENS.

Since the specific resistance is inversely proportional to the square of the radius, it is necessary that the radius should be accurately known: and if the specimen is only a few millimetres in diameter it is difficult to measure it with great accuracy. In the present experiments the specimens used were thin, the actual dimensions being 9 centimetres long by 3 millimetres diameter. It was concluded, after considering all the factors which could be foreseen, that the advantages of thin specimens far outweighed their disadvantage. For example, if the diameter is large, a satisfactory potential drop across the knife edges can only be secured by passing a current of several amperes through the specimen. Unless a very large battery of accumulators is available, such a high current will give rise to inconvenient fluctuations. This difficulty could, of course, be overcome by using very long specimens, but in practice it is essential that the specimens be comparatively short, as otherwise they are liable to bend on quenching. An even more important point is that thick specimens readily crack on quenching. Edwards (35) used specimens 1.1 centimetres in diameter when investigating the resistance of chromium steels. The majority of the specimens cracked, and he was obliged to take the minimum diameter of the bar, in the plane of the crack, as the diameter of the specimen.

None of the specimens used in the present investigation cracked or bent on quenching. The method by means of which the cross-sectional area of the bars was measured with satisfactory accuracy, will be described later.

For these resistance experiments 4 specimens were made from each of the 19 steels used in this work. Some difficulty was experienced in the manufacture of these 76 specimens, which, because of their length and thinness presented some difficulty

difficulty in machining. They were turned down from 1 inch round bar to approximately the correct diameter and finished by grinding.

All the specimens were annealed at 850° C. and the annealed resistance of each was determined. They were then quenched from different temperatures and the resistance again measured.

APPARATUS.

Two potentiometers were used. On one, the drop in volts across 5 centimetres of the standard and across 5 centimetres of the unknown specimen was measured alternately. On the other, simultaneous measurements of the current were made. The current was supplied from a battery of 16 Hart accumulators and was regulated by means of two adjustable rheostats. By means of a standard shunt the current was measured on a potentiometer. Two pairs of fixed knife edges were used, one pair for the unknown specimen and one pair for the standard. These knife edges were made of nickel-chromium steel. They were soldered to a brass plate which in turn was screwed to a wooden slab. The knife edges were therefore parallel and the distance between them was invariable. Heavy rubber bands were used to press the specimens firmly on the knife edges.

Arrangements were made so that the current could be switched through an alternative circuit when readings were not being taken. This was done because it was found that the current fluctuations were more serious during the first few minutes after switching on. Further, as prolonged passage of the current through the specimens would raise the temperature slightly, it was thought advisable to provide this alternative circuit.

MEASUREMENT OF THE RESISTANCE OF THE STANDARD SPECIMEN.

The first step in the investigation was the measurement at 17°C. of the resistance of the annealed specimen which was to be used as a standard during the experiments. For this purpose the specimen was placed on one pair of fixed knife edges and held there by means of the rubber bands. A Beckmann thermometer was placed in contact with the specimen, which was enclosed in a box so that its temperature could be accurately measured. A current of 1.5 amperes was allowed to flow through the circuit, and the potential drop between the knife edges was measured at intervals. Simultaneously with each measurement the current flowing was determined by means of the shunt and the other potentiometer. The resistance of the specimen at the temperature of the experiment was obtained by calculation.

Readings of resistance were taken during several days, the temperature of the specimen at the time of each reading being read to one hundredth of a degree. The results of the standardisation were plotted, and the resistance of the standard specimen at 17°C. was obtained from the curve.

MEASUREMENT OF THE RESISTANCE OF THE HEAT-TREATED SPECIMENS.

The heat-treated specimen, the resistance of which was required, was firmly fixed to the other pair of knife edges and joined in series with the standard. The current, which had been flowing through the alternative circuit till fairly constant, was switched through the specimens, and the potential drop across 5 centimetres of the standard was read, a simultaneous reading of the current being made. The potential drop across the unknown specimen was then measured, the current being measured as before. Immediately this reading had been taken the current and potential drop across the standard was again measured, and the current switched back to the alternative circuit.

On completion of this group of readings, the specimen was moved several millimetres on the knife edges, and after a short interval of time a series of readings was taken as before.

Several groups of readings were taken for each specimen, the number depending on the uniformity of the steel. From each group the resistance of the specimen at 17°C. was calculated, and the mean of these values was taken as the resistance of the specimen.

In order to find the specific resistance of the steel, it was necessary to find the mean cross-sectional area of each specimen. This was done by measuring the diameter at equal intervals along and around the specimen.

The specimen was held by elastic bands between two grooved supports on a wooden base. The length of the base between the supports was divided into twelve equal parts. By means of a micrometer, readings of the diameter were taken at points on the specimen directly above each of these divisions on the base scale. By means of a pointer fixed to one end of the specimen, this was rotated through 60°, and the diameter again measured at twelve equidistant points. A final set of readings was taken after rotating the specimen through 120°. The quarter square of each diameter was found and the mean of these 36 quarter squares was taken as the square of the radius.

It should be noted that the specific resistance is inversely proportional to the square of the radius, and this means that, not only should the diameter be measured accurately, but the cross-sectional area should be calculated at each measured diameter, and the average of these values found. The cross-sectional area is, of course, proportional to the square of the radius.

This somewhat elaborate method of determining the cross-sectional area of the specimens was found to be necessary.

Repeated attempts to make the measurement by the ordinary casual method showed that a slightly different value could be obtained each time. By the method described above it was possible to repeat the measurements time and again with absolutely identical results.

In tables 14 to 54 the results of the electrical resistance measurements are shown in detail. Column 1 contains the resistance as measured, column 2 contains the mean resistance, column 3 contains the mean square of the radius, and column 4 contains the specific resistance. In tables 55 to 58 these results are summarised. The quenched results are the specific resistances from tables 14 to 54, but the annealed results are the mean of the four specific resistances obtained for the four different specimens of each steel.

The specific resistances in these tables and as plotted in figures 19 to 24 are as obtained. They have not been corrected for manganese, silicon or phosphorous. As some of the results, with which these present results will be compared, are corrected, the corrections are shown in table 13.

Before proceeding to the consideration of these results, a point which is frequently overlooked in resistance determinations may be mentioned. When the specimens as quenched are martensitic, the diameter will be slightly larger than in the annealed state. It is therefore not strictly correct to take the diameter as being constant during heat-treatment. In the present work the diameter of each specimen was measured in both the annealed and quenched states.

SECTION 3. SUB-SECTION C.

ELECTRICAL RESISTANCE DETERMINATIONS: CONSIDERATION OF RESULTS.

ELECTRICAL RESISTANCE. CONSIDERATION OF RESULTS.

On the whole the specific resistance determinations do not reveal anything unexpected. Except for certain minor points the results are exactly what would be expected from the iron-carbon diagram.

The general features of the effect of carbon on the resistance of iron have long been established, but as they will require to be borne in mind throughout the present discussion, they may be re-stated briefly.

Cementite has a greater resistance than alpha iron.
An amount of carbon in solution in either alpha or gamma iron, raises the resistance of the iron to a greater extent than a similar amount of carbon as cementite.
Nickel and Chromium both raise the resistance of iron.

Quenching should therefore have the following effect on the specific resistance of steel:- By retaining more carbon in solution than is retained by slow cooling, it should raise the resistance. Increase of the rate of cooling, either by raising the quenching temperature or by changing the quenching medium, should increase the resistance by increasing the amount of carbon retained in solution. In hyper-eutectoid steels increase in the quenching temperature should increase the resistance by bringing more of the pro-eutectoid cementite into solution.

The results are in agreement with the foregoing account of the effect of heat treatment. In Figures 19 to 22, it will be seen that the resistance of hyper-eutectoid steels is increased by raising the quenching temperature, whereas the resistance of hypo-eutectoid steels is decreased. Why the resistance of these latter steels should be affected in this way by quenching temperature, is not directly explained by the diagram. Other investigators have found that the resistance of hypo-eutectoid steels is

is increased by raising the quenching temperature. This effect, found by Campbell (36) and Saldau (37), is not directly explained by the diagram either. It demands some other explanation. It may be due to changes taking place at temperatures above the ordinary critical points, as these authors have suggested: but probably the cause is simpler.

McCance (38), Le Chatelier (39) and Benedicks (40) found no increase in the specific resistance of hypo-eutectoid steels when the quenching temperature was raised. This is what would be expected from the diagram.

A decrease in specific resistance, such as was found in the present investigation, is more in accordance with the equilibrium diagram than the increase noted by some previous authors. For in steels with less than .9 per cent of carbon, although the carbon is all dissolved at the Ac1, further heating does promote diffusion of the carbon and the attainment of a homogeneous state. At the lower quenching temperatures the steel consists of areas relatively rich in carbon, and areas of comparatively pure ferrite. At the higher temperatures a uniform solid solution exists. This change from heterogeneity to homogeneity may have no effect on the specific resistance of the quenched specimens. On the other hand it may lower it, but it is not likely to cause an increase.

It will be seen in Figure 20 that increasing the quenching temperature from 1000° C. to 1100° C., has practically no effect on the specific resistance of the steel containing 1.61 per cent of carbon. Yet a similar variation in the quenching temperature has a pronounced effect on the specific volume and electrode potential. Evidently the specific resistance is only influenced by the amount of carbon in solution, whereas specific volume and electrode potential are sensitive to the distribution of this carbon, and to the relative amounts of martensite and austenite produced by quenching.

Further evidence of this insusceptibility of resistance measurements to changes other than alteration in the amount of carbon in solution, is the fact that a straight line can be drawn through the specific resistance of carbon steels quenched from 1100° C. When it is remembered how the nature and relative amounts of the constituents vary from end to end of this series, it is surprising that specific resistance and carbon content should be directly proportional. Resistance apparently measures the average amount of carbon in solution in the steel.

The line connecting the specific resistance of annealed carbon steels with composition, shows a break at .9 per cent of carbon. Above this amount of carbon the resistance increases less rapidly with the carbon content. Apparently the amount of carbon dissolved in the alpha iron increases with carbon content to .9 per cent. Further additions of carbon produce free cementite only.

The specific volume and electrode potential results also indicate this increased solubility of carbon in alpha iron. And in the section devoted to discussion of the electrode potential results, it is suggested that this phenomenon arises from the fact that it is not pure iron which separates from austenite, but ferrite of variable composition. The electrode potential results show that the amount of carbon dissolved in the first area formed, increases with the carbon content. The specific volume and electrical resistance results show that the total amount of carbon dissolved in the ferrite increases with the carbon content. All this points to the similarity between the formation of the solid solution ferrite from the solid solution austenite, and the formation of a solid solution from a liquid solution. It is not possible from these results to determine the position of this "solidus", but it evidently departs from the vertical to a greater entent than is commonly believed. It is doubtful whether the exact position of this "solidus" is of any importance. Its position will depend so greatly on the

the rate of cooling, on the amount of impurities, on the previous heat-treatment of the steel, and on the degree of homogeneity produced in the austenite, that any experimentally determined line would only have a real existence under the exact conditions which prevailed during the experiments. The presence of this line cannot, however, be ignored. It undoubtedly plays an important part in determining the properties of ferrite and pearlite produced under various conditions of cooling, in determining the conditions necessary to produce sorbite and primary troostite, and in giving to these constituents their characteristic features.

In another section evidence will be adduced to show that in pearlite, troostite and sorbite, the carbon content of the ferrite changes progressively with change in the state of division This is in agreement with the view that sorbite and primary troostite are produced, when the Ar1 takes place in such a way that considerable crystal growth around a small number of nuclei gives place to a small amount of growth round many nuclei, and ultimately to the 'labile'shower' or simultaneous formation of nuclei throughout the metal. This change in the nature of the Ar1 is, of course, accomplished by increasing the rate of cooling.

All other investigators have not obtained the same results for annealed carbon steels. Matthiessen (41) concluded that the specific resistances of annealed carbon steels lay on a straight line when plotted against carbon content. Le Chatelier arrived at the same conclusion. But Gumlich (42) found a break in the linear relationship at .9 per cent of carbon. It is probable that this diversity is caused by the difficulty of deciding what the annealed state is, rather than by inaccurate methods of measuring the resistance.

The specific resistance of carbon steels after tempering for 6 hours at 240° C. is shown in Figure 20. It is clear that some carbon still remains in solution, though the tempering was

was practically complete at that temperature. This point will be discussed at some length in the section dealing with the continuous change of resistance during tempering.

When the values for the specific resistance of carbon steels quenched from 1100°C. are corrected for silicon, manganese and phosphorus, according to the figures given in Table 13, the line through the corrected values may be expressed by the equation:

$$R = 7.5 + 26.7 \text{ Carbon } \%.$$

This agrees perfectly with Benedicks' formula:

$$R = 7.6 + 26.8 \text{ Carbon } \%.$$

Benedicks' formula, however, is supposed to be applicable to all steels, and instead of carbon percentage, the summation of the carbon equivalents of all elements, other than iron, is used. This wider application of the equation does not receive support from the resistance results in the present thesis. As far as the special steels used in this research are typical of the class, the formula does not meet the facts.

From the above formula it is evident that the specific resistance of pure iron should be 7.5 microhms. This question of determining the specific resistance of pure iron by extrapolating the lines through a series of steels has already been discussed, and it was said that the line through the corrected resistances of annealed carbon steels met the resistance axis at 9.5 microhms, which agrees with the values for the specific resistance of pure iron determined by Guillet (43), Gumlich (44) and Yensen (45).

This discrepancy between the values determined by extrapolation, is undoubtedly due to the presence of carbon in solution in the ferrite of annealed steels. If annealed steels were simply mixtures of pure iron and cementite, the resistance would be lower, and the line through the specific resistance of a series

series of annealed steels would probably intercept the resistance axis in the vicinity of 7.5 microhms. The specific resistance of pure iron is therefore about 7.5 microhms as suggested by Benedicks and confirmed by Thompson.

In Figure 24 the corrected specific resistance of all annealed steels is shown. The curves are similar: in all cases the slope is less steep in the hyper-eutectoid range.

In Figure 23 the corrected specific resistance of all steels quenched from 1100°C. is shown. The curves for the nickel and chromium steels almost coincide. When this is compared with the relative position of the resistance curves of these steels in the annealed state, it is evident that there is more chromium in solution in the quenched than in the annealed state, whereas nickel remains in solution in all conditions.

In every case the curves for nickel-chromium steels lie above all others. It can be shown that in a nickel-chromium steel of a given carbon content, the resistance is increased above that of a corresponding carbon steel, by an amount that equals the sum of the increase due to each of the special elements.

Thus, in steels containing .51 per cent of carbon, taking the actual resistance of the carbon, chromium and nickel steels from the curves, and allowing for the difference in the percentage of nickel and chromium in the nickel-chrome steel, the calculated resistance of the latter is 37.5 microhms. The experimentally determined specific resistance of this steel is 37.6 microhms. Calculating in the same way, the specific resistance of a nickel-chromium steel containing 1 per cent of carbon is found to be 49.0 microhms. The value actually determined was 48.0 microhms.

In the course of the present work it was found that the effect of special elements on the resistance of steel, depends on whether the steel is quenched or annealed, on the quenching temperature, and on the carbon content.

Thus the addition of 1 per cent of chromium increases the re-

the resistance of quenched steels by about 4.6 microhms, when the carbon is under 1 per cent. Above this percentage of carbon the effect of chromium becomes gradually less, until at 1.6 per cent of carbon the increase in resistance produced by the addition of 1 per cent of chromium is only 3.3 microhms.

In the annealed steels the effect of chromium is considerably less than this, and varies noticeably with the carbon content. According to Benedicks' formula, the increase in the specific resistance of iron due to each unit per cent of chromium, should be 6.2 microhms. The results given above show that the effect of chromium cannot be so simply expressed.

It is the same with nickel steels. In low carbon steels 1 per cent of nickel increases the specific resistance by 2.3 microhms. In high carbon steels the effect is somewhat less - about 1.9 microhms. This applies to the quenched and annealed states equally.

According to Benedicks' formula, 1 per cent of nickel should increase the resistance by 5.4 microhms.

Little need be said about the curves for the special steels, Figures (19-22). Increasing the quenching temperature does not increase the resistance except in the high carbon steels. As in carbon steels there is some indication of a decrease in the specific resistance with higher quenching temperature.

The results obtained for the chrome steel, containing 1.73 per cent of carbon, quenched from 1100° and 1000° C. are peculiar. The higher quenching temperature gives the lower resistance. Quenched from 1100° C. the steel is almost entirely austenitic, whilst from 1000° C. it is largely martensitic. The most obvious explanation is that austenite has a lower specific resistance than martensite. This fact, however, does not agree very well with the carbon and nickel-chromium steels. On the other hand, a nickel steel containing 1 per cent of carbon gave its maximum specific resistance when quenched from 900° C.

Quenched from higher temperatures the specific resistance was considerably lower. As this steel did contain more austenite when quenched from the higher temperatures, it seems reasonable to attribute this fall in specific resistance to the appearance of considerable quantities of austenite. To throw light on this point, a number of high carbon special steels was quenched from 1100° C. and subsequently immersed in liquid air for an hour. The specific resistance was measured before and after immersion, with the following results:-

Chrome steel 1.73% carbon. Resistance fell .4 microhms.
Nickel-Chrome steel 1.28% carbon. Resistance fell 1.1 microhms.
" " " 1.51% carbon Resistance fell 3.0 microhms.

Specific volume measurements showed that austenite had been largely converted to martensite during immersion in the liquid air. According to these results, austenite has a higher specific resistance than martensite.

During determination of the tempering resistance curves, described in another section, evidence was obtained to the effect that at 240° C. martensite has a higher resistance than austenite. Perhaps this question might well be left at this. Austenite and martensite are not definite physical entities. They vary in properties as they vary in constitution, and although they are distinguishable by means of certain physical properties, this is not always so.

The values obtained for the chrome steels tempered at 240° C. are interesting. If they are compared with the carbon steels similarly treated, it will be seen that the difference between the chrome and carbon steels as tempered, is greater than the difference between then as annealed or as quenched. This means that the chromium has remained in solution during tempering and that, due to the presence of this chromium, the separation of carbon has been retarded.

Therefore for chromium steels, tempered at 240° C., the

the limiting state is much farther from the annealed state than it is for carbon steels tempered at the same temperature. Whether the tempering which does take place is accompanied by separation of carbon only, or of chromium and carbon together, cannot be ascertained from the available results.

In concluding this section it is appropriate to refer to some observations made during the quenching of these resistance specimens. Although the specimens were very thin and were quenched rapidly in iced brine, cooling was not instantaneous. It was therefore possible to observe that the ends of the specimens became black while the middle portions were still red. This indicates that the rate of transference of heat along the specimen is notably less than the rate of transference of heat from the specimen to the bath. This fact is commonly recognised, and would not be referred to here had it never been questioned. But Campbell[46] endeavoured to show that the contrary held. He argued that when a long specimen is quenched by lowering it into the bath endwise, the last end to enter the water receives the most effective quenching. He explained this reasoning by saying that the cooling of the first end is retarded by conduction of heat along the specimen from the the portion not immersed, so that this first end cools more slowly than the end last to enter, for the end that enters last cannot receive heat from the first end, which is already cold.

There is no doubt that the conditions which pertain during the cooling of differently shaped specimens of materials of different thermal conductivity in media of different cooling power, have received insufficient attention from metallurgists. The writer is convinced that a thorough investigation of these things from the standpoint of mathematical physics, would be of great assistance in giving a quantitative meaning to many conceptions now held qualitatively. It is, in fact, not improbable that nearly all those problems peculiar to metallurgists that

that have their origin in the nature of the materials dealt with, will be ultimately solved by the application of methods of precise calculation to the phenomena of cooling, and diffusion in the solid state. Qualitative knowledge has its usefulness, but only within limits. It lends itself too readily to manipulation, its significance is too subjective, and in consequence, the image of a phenomenon known only qualitatively is impressed differently on different minds, and distortion of any feature or aspect of the image can only be distinguished as distortion, when it exceeds by a considerable margin the currently held image of the objective reality.

Campbell's reasoning is undoubtedly fallacious. It exceeds the limits of tolerance permitted by the current conception of cooling phenomena.

It is scarcely worth while to examine his reasoning more closely, but a few remarks may be made on the experimental method by which he verified his conclusions. He used the method of thermo-electro-motive force. He had previously shown that the electro-motive force generated at a junction between an annealed bar and a quenched bar, increased as the quenching was made more drastic. Bars 15 cms. long were then quenched endwise and the electro-motive force of each end was measured. It was found that the couple formed by the end which entered the quenching bath last and the annealed bar, gave a greater E.M.F. than the couple formed by the other end and the annealed bar. This was taken to show that the end which entered last had been the most effectively quenched.

The explanation of this anomalous discovery is to be found in the experimental method used. The bars forming the couple were disposed so that they formed a wide V. The apex of the V was immersed 8 cms. below the surface of ~~the surface of~~ the oil bath used. Therefore considerably more than half the length (15 cms.) of the bar was actually in the oil. With an arrangement

arrangement of this kind, the E.M.F. measured would be that between a point on the quenched bar at the surface of the oil, and the annealed bar. In short, the E.M.F. measured was that of a point near the end of the specimen farthest from the junction and not that of the end, which with the annealed bar formed the junction.

SECTION 4. SUB-SECTION A.

CONTINUOUS MEASUREMENT OF ELECTRICAL RESISTANCE DURING TEMPERING.

EXPERIMENTAL.

MEASUREMENT OF RESISTANCE DURING TEMPERING.

In this section a method of measuring continuously the change in resistance during tempering is described.

Curves showing the change of resistance of 19 steels, as quenched from different temperatures, are given and the significance of these curves is discussed.

PRELIMINARY REMARKS.

Before proceeding to the description of the method by means of which continuous observation of the tempering process was accomplished, some account of the considerations which inspired these experiments may not be out of place.

The customary means of investigating the tempering process consists of heating intermittently at the selected temperature, and making the required observations or measurements at ordinary temperature in the intervals of heating. Now this procedure is not strictly correct. It involves a heating up and a cooling down between each observation. In accordance, however, with that view of the tempering process which has attained, if not general acceptance, at least considerable credence, the objection above raised is habitually passed over as a fact of minor importance. It is held that the only effect of this heating and cooling is to add a little to the amount of tempering which takes place. There is therefore no reason, according to this view, why continuous measurement at the actual temperature should reveal anything not made evident by the alternative method. It is all so clear. The specimen is to be heated for 3 hours at 200° C.: surely the tempering which takes place during the few minutes heating to and cooling from this temperature may safely be neglected.

The specific volume results included in this thesis were obtained by the customary method. Some points about these results seemed rather inexplicable, so, as the resistance is

is a property which lends itself to continuous measurement, it was decided to perform the experiments on the change of resistance during tempering, in such a manner as would avoid the necessity for intermittent heating and cooling. The object was to ascertain whether there was any basis for the aforesaid objection or not.

The apparatus, by means of which continuous observation of the tempering process was carried out, was designed to surmount a number of difficulties which are inherent in experiments of this nature. These difficulties were not appreciated when the project was first conceived, but during a number of preliminary experiments they became sufficiently evident. Two principal difficulties may be specifically referred to.

In the first place the temperature must be maintained within very narrow limits during the tempering, otherwise changes in resistance due to tempering may be obscured by changes due to temperature fluctuations.

Secondly, as the tempering of martensite proceeds at first with great rapidity, provision must be made for the taking of readings immediately after the commencement of tempering, and at very short time intervals thereafter.

In addition to these two principal difficulties, the preliminary experiments directed attention to a number of subsidiary points, such as the prevention of oil films on the electrical contacts, which the apparatus used would take into account.

DESCRIPTION OF THE APPARATUS AND METHOD EMPLOYED IN THE CONTINUOUS MEASUREMENT OF RESISTANCE DURING TEMPERING.

The general layout of the apparatus is shown in figure 25 and the details of the tempering box in figure 26.

For descriptive purposes the apparatus may most conveniently be considered in three sections:

(1) The Oil Heating and Circulating System.

(2) The Current Supply System.

(3) The Apparatus for Measuring the Resistance.

THE OIL HEATING AND CIRCULATING SYSTEM.

A copper tank, containing about four gallons of heavy cylinder oil, was heated electrically by means of nichrome elements and resistance mats. By means of an ordinary sud pump the oil was drawn from this tank and pumped into a cast brass box of small capacity, in which the specimens were suspended. A baffle plate provided equal distribution of the incoming oil. From the other end of the tempering box the oil flowed under gravity back to the heating tank.

Two heavy brass lids were provided for the tempering box. The under surface of these lids was machined so that they fitted closely on the faced flange of the box. Four bolts with wing nuts were used to hold the lid firmly in position. One of these lids was for use during the time that the oil was heating up and until the correct experimental conditions were realised. When all was ready this lid, which carried a thermometer only, was removed, and the other lid which carried the specimens was placed in position.

Each specimen required four contacts, two knife edges for measuring the fall of potential over a known length, and two other connections by means of which the desired current could be passed through the specimen. Provision was made for three specimens besides the thermometer. The method of holding these specimens on the lid was as follows:

A hole $5/8$" diameter was bored in the brass lid at the point where each contact was to be made. On top of this brass lid a uralite lid was placed. The holes in the uralite corresponded with those in the brass, but were of smaller diameter. Three strips of heat resisting, insulating red fibre with small diameter holes were placed on the uralite. Three inch lengths

lengths of screwed brass ~~brass~~ rod were then passed through the whole in such a way that they were gripped by the uralite and fibre, but passed with considerable clearance through the brass. The space between the rod and the brass lid was tightly packed with asbestos to ensure that no short circuit between the rod and the brass would occur. Strips of the insulating fibre already mentioned, were placed on the underside of the lid so that the rods projected through them. By means of nuts on the rods the whole was tightly screwed up. This arrangement provided perfectly insulated contacts through the lid.

The ends of the rods which projected from the top side of the lid acted as the terminals for the electrical connections. To the other end of the rods the knife edges and current connections were attached by means of thin brass strips. These knife edges were made of nickel chromium steel, and a bolt passing through each pair enabled them to be clamped tightly to the specimen. The method used for passing the current through the specimens was designed to prevent the oil from penetrating between the specimen and the contact. A thin strip of lead foil was first wound round the specimen. Two notched pieces of brass were then placed so that the lead foil came between the notches. By means of a bolt through them, the two brass pieces were pressed very tightly together, and the specimen was gripped in the lead foil.

Each time a new specimen was placed on the lid all the connections were taken to pieces and carefully cleaned with benzene and emery.

To minimise heat losses the heating tank and the tempering box were packed in asbestos wool. The system of pipes was heavily lagged.

During the experiments the temperature was controlled by regulating the current flowing through the heating elements and mats. It was found by experiment that the temperature at

at different positions in the tempering box did not differ perceptibly at any time.

THE CURRENT SUPPLY SYSTEM.

Although it was not found possible to get a source of current which was perfectly invariable, a battery of twelve 1.3 volt "Nife" cells proved very consistent. The current from this battery passed through an ammeter used for rough measurement, through a 0.1 standard shunt used for accurate measurement, through two rheostats for adjusting the current to the desired value, and then to the specimens. The specimens on the lid were connected in series. Accurate measurement of the current was obtained by reading on a potentiometer the drop in volts across the shunt.

THE ELECTRICAL MEASURING APPARATUS.

The electrical resistance was determined by measuring the drop in volts across a known length of the specimen, when a current of a given value was flowing. Four measurements were required, the drop in volts across each of the two specimens, the standard specimen and the standard shunt. Two Tinsley Vernier Potentiometers were used to obtain these measurements, and full details of the connections are shown in figure 25.

Although a distance piece was used to fix the length of specimen between the knife edges, it was not found possible to carry this out with accuracy. The difficulty was overcome in the following manner:

The specific resistance of the specimen at 15°C. was determined between the fixed knife edges as described under Electrical Resistance. When the specimen was attached to the lid its resistance at 15°C. was again measured. From the specific resistance, the resistance between the knife edges and the cross sectional area, the exact length of specimen between the knife edges on the lid was calculated.

DETERMINATION OF THE RESISTANCE OF THE STANDARD SPECIMEN.

It was absolutely indispensable, to the satisfactory performance of these tempering experiments, that a standard specimen be used. This specimen whose resistance at 240° C. was accurately known, was kept permanently fixed in its position on the lid. By taking simultaneous measurements of the potential difference across the knife edges on the standard specimen and the unknown, small variations in temperature and conditions were accounted for. At the same time the standard served to preserve the relation between the results obtained from day to day.

An annealed nickel steel was used as the standard.

The first step in the investigation was to determine the resistance of this nickel steel specimen at the tempering temperature which was conveniently fixed at 240° C., being the temperature at which the apparatus attained its greatest constancy. It was originally intended to perform the experiments at a somewhat higher temperature, but because of the ease with which 240° C. could be maintained, it was decided wo work at that heat.

The standard specimen was fixed between the knife edges on the lid. Each end was wrapped in lead foil and tightly gripped between the clamps to which the current connections were attached. When the apparatus had attained the desired temperature, the lid carrying the standard specimen was placed in position on the tempering box. A large number of simultaneous readings of the current and the voltage drop between the knife edges was taken, and the resistance calculated. During these determinations, the temperature of the oil, which was measured by a mercury thermometer, was varied from several degrees above 240° C.to several degrees below. This procedure was repeated for some days in order that any change which might occur in the specimen during prolonged soaking at this temperature, would complete

complete itself.

At the conclusion of these experiments all the resistance values obtained were plotted against temperature, and the true value at 240° C. read from the graph.

Before the standard was placed on the lid, its resistance was measured on the fixed knife edges used in the ordinary resistance determinations. It was therefore possible to connect the resistances measured at 240°C. with the resistances measured in the cold.

PROCEDURE DURING THE EXPERIMENTS.

Although the whole apparatus was thoroughly lagged, it required about five hours to heat up to 240° C. and to become steady at that temperature. So that work could be started at 9.30 a.m. and a whole day be available for tempering, it was necessary that heating should begin at 4.30 a.m. Moreover, the cylinder oil was so viscous at low temperature that the circulating system could not be started in the cold. The following arrangements were therefore made for starting:

At 4.30 a.m. an alarm clock released a rat trap which pulled the main switch in the heating circuit. At the same time, it pulled the catch on another clock into the alarm position. This second alarm was set for 6.30 a.m. and at that time it switched on the current for the oil pump motor. By 9.30 a.m. all preparations for the experiments had been made, and the current, which had been running for half an hour, had reached satisfactory constancy after the usual initial period of fluctuation. Since the current factor was eliminated by taking simultaneous readings of the volt drop across the standard and the quenched specimen, no readings of the current were taken during the tempering.

When the current had attained constancy, the resistance at 15° C. of the specimen as fixed on the lid, was measured for

for the purpose of calculating the exact length between the knife edges.

The lid was now placed in position on the box so that the specimens were immersed in the oil, and readings of the voltage drop across each of the specimens were taken one half minute after immersion, at intervals of half a minute thereafter up to ten minutes, at intervals of one minute till twenty minutes, and as frequently thereafter as seemed desirable.

Although readings were taken only at the intervals stated, the spots of light on the galvanometer scales were watched continuously. An infinite number of readings could equally well have been taken, but having regard to the elaborate calculations involved, it was considered sufficient to take readings at the intervals mentioned.

The specific resistance of the tempering specimen at 240° C. was calculated at each reading by means of the formula:

$$Rt' = \frac{E't \times R \times a}{Et \times l}$$

where Rt' = specific resistance of the tempering specimen at time t.

Et' = volt drop across length of tempering specimen between knife edges.

Et = volt drop across knife edges on standard specimen.

R = resistance of the standard at 240° C.

a = cross sectional area of tempering specimen as determined by method described under electrical resistance.

l = length of specimen as calculated from its cross sectional area, resistance between fixed knife edges at 15° C., and resistance between knife edges on tempering box at 15° C.

As will be seen from the formula, the true specific resistance of the specimen at 240° C. is obtained if it is assumed that the temperature coefficient of the standard is the same as that of a hardened specimen. This assumption is, of course, not

not perfectly correct; but as the temperature of the oil bath did not, at any time after immersion, differ from 240°C. by more than ± 2°C., the error introduced is neglible.

After withdrawal from the oil, the specific resistance of each tempered specimen at 15°C. was measured between the fixed knife edges.

DETERMINATION OF RESISTANCE AT ZERO TIME.

When the results obtained during the measurement of the change of resistance at 240°C. were considered, it was evident that the fall in resistance during the first few minutes was very great. During the first minute the fall would be greater than in any subsequent minute, and the fall during the first minute was unknown. In most cases the first reading had been obtained after one half minute, in exceptional cases after one quarter minute, and in other instances not until one minute after immersion in the oil. This, of course, resulted from the difficulty of balancing an unknown resistance on the potentiometer. Before immersing the specimens, the potentiometers were set at the anticipated readings. While this could be done with reasonable accuracy for the standard specimen, it was largely dependent on pure probability whether the reading for the unknown specimen would be satisfactorily guessed.

Immediately after immersion the resistance of the unknown and the standard specimen rose rapidly, and the first reading was therefore dependent on how quickly these progressively changing values could be balanced. Because of their small diameter the specimens very quickly attained the temperature of the oil, and thereafter readings could be taken with some accuracy. But the first reading was only accurate so far as the temperature coefficient of the standard specimen equalled that of the unknown. Therefore the first readings taken were not obtainable at any definite time after commencement of tempering, and were also unreliable in so far as the temperature which the specimen

specimen had attained at the time of taking the reading, differed from the true tempering temperature.

In order to complete the curve it was necessary that the reading at zero time at 240° C. should be known. Because of the steepness of the curve at this point, this could not be properly accomplished by extrapolation. The only available means of determining the starting point of the curve was to measure the temperature coefficient of resistance of each specimen. The specimens were therefore re-quenched from the same temperature as before, and their temperature coefficient measured between 0° C. and 80° C. For this purpose, the specimens were immersed in a water bath, the temperature of which was gradually raised by the addition of boiling water. Readings of resistance were taken every few degrees, the temperature being allowed to become steady before each measurement. The values were plotted, and the temperature coefficient of resistance was calculated from the line through the points. From these coefficients, which are given in tables 99 and 100, the specific resistance of the quenched specimens at 240° C. was calculated, and the values so obtained were taken as the starting points of the curves.

These tempering resistance curves are given in tables 59 to 98 and are plotted in figures 27 to 39.

Before proceeding to a detailed consideration of the curves obtained by the foregoing method of continuous measurement of resistance, it may not be out of place to anticipate here some conclusions which do not actually appear until the end of the consideration. These conclusions bear principally on the experimental work and are therefore appropriate here. It becomes apparent from consideration of the curves obtained, that there is no appreciable difference in the curves for the same steel quenched from different temperatures. Again, during the mathematical treatment of the curves, it is shown that all steels

steels in a particular series give rise to curves which are suitably represented as the resultant of three general curves. It is thus clear that one steel of a moderate carbon content, which when quenched consists of martensite only, would be sufficient to give those general components for any particular series. Unfortunately it would not have been permissible, even had it been possible, to anticipate these conclusions before performing the experimental work, but now that the experimental work has been accomplished, it is certainly important that the significance of this should be realised in order that subsequent work may be simplified.

SECTION 4. SUB-SECTION B.

CONTINUOUS MEASUREMENT OF ELECTRICAL RESISTANCE DURING TEMPERING.

CONSIDERATION OF RESULTS.

TEMPERING RESISTANCE EXPERIMENTS. CONSIDERATION OF RESULTS.

The foregoing method of making continuous measurement of the change in resistance during tempering was used to determine the tempering changes of all the steels in table 1. It was originally intended to investigate the tempering changes of each steel as quenched from 800°, 900°, 1000°, and 1100° C., but at an early stage in the work it became evident that there was no difference in the curves for different quenching temperatures. Consequently the quenched 800° C. and 900° C. series are incomplete. In tables 59 to 98 the time-resistance figures for each curve are shown, and in figures 27 to 39 the curves of resistance against time are plotted in suitable groups.

In certain of the curves, noticeably those of the medium carbon steels, it will be seen that the greater part of the tempering, as measured by the fall of resistance, takes place in a very short time after commencement of tempering. In the steels with higher percentages of carbon, particularly those of the chromium and nickel-chromium series, the fall of resistance is conspicuously slower. It appears that those steels in which martensite is the predominant constituent temper very rapidly, whereas those in which austenite predominates temper much more slowly.

It is a reasonable deduction from these curves that martensite breaks down into troostite or sorbite with great rapidity, and that austenite tempers more gradually, only a small proportion of the total amount of that constituent decomposing at any given time. The process is as follows:-

A small amount of austenite breaks down into martensite, which immediately decomposes into ferrite and free cementite and is replaced by further decomposition of austenite. This process continues till all the austenite has tempered. It gives rise to a slow steady fall of resistance such as is seen

seen in the curve of the chrome steel containing 1.73 per cent of carbon quenched at 1000°C.

If at any time during the tempering of a quenched steel the austenite suddenly decomposes into martensite, the occurrence will be followed immediately by a rapid decomposition of this martensite. No evidence of this was observed in any of the curves.

It should be pointed out that although readings were only taken at the time intervals stated, the fall of resistance was actually watched at all times during the tempering by following the movement of the spot of light on the scale. Consequently no sudden change in the rate of tempering could have been overlooked. The fact that martensite tempers much more rapidly than austenite shows that there could not possibly be an accumulation of martensite during tempering of austenite at 240°C.

Reference to figure 1 will show on tempering at 250°C. there was so great an accumulation of martensite, formed by the decomposition of austenite, that in a high carbon chromium steel the specific volume suddenly jumped from the very low value of austenite to the high martensitic value. The same effect was noticed in a lesser degree in chromium steels of lower carbon content. In the case of the steel containing 1.73 per cent of carbon, it was evident that on tempering for three hours at 250°C. the bulk of the austenite had broken down with the formation of martensite, which was still present, undecomposed, when the specific volume was determined. In other words, on tempering at 250°C. there was a large accumulation of martensite.

This fact appears to contradict the statement that no accumulation of martensite ever occurred in a tempering austenitic steel at 240°C. It should be noted, however, that the specific volume was not determined at 250°C., but that the specimens were cooled down from that temperature to 15°C. at which temperature the specific volume was measured. The decomposition of

of the large amount of austenite and the accumulation of martensite must have taken place during the cooling.

This suggestion is confirmed by the curves shown in figures 38 and 39. From these it will be seen that, while austenitic steels were tempering slowly and steadily, the specimens were removed from the tempering bath and allowed to cool down to room temperature. When the specimens were replaced in the tempering bath, it was found that the resistance had risen several microhms due to accumulation of martensite, which now tempered rapidly in the characteristic manner.

In figures 83 to 86 are shown the dilatation curves of some of the steels in the nickel-chrome-molybdenum series. These curves were taken with specimens which had been oil quenched from 1000° C. and which were composed, to some extent, of martensite and austenite. During the tempering of these steels there was no sudden change of any kind. In most cases a slight change of slope is the only anomaly observed. Curve 1, figure 86, represents the tempering of steel No.5 after quenching from 1000° C., and although this steel did most certainly contain a considerable quantity of austenite, there is no evidence of a sudden expansion due to austenite changing to martensite, nor of a rapid contraction due to martensite changing to troostite.

Enlund (4) comments on this point. When making specific volume determinations on a series of steels tempered at progressive rising temperatures, he observed a pronounced increase in specific volume after tempering between 200° C. and 300° C. He remarks that such an expansion has never been observed during dilatation experiments on steels containing austenite, and implies that the change has escaped notice. This same author measured the resistance of quenched steels continuously during progressive heating, and shows temperature-resistance curves for a number of steels. All these curves show two anomalies, one between 110° C. and 120° C. and another between 250° C. and

and 260° C. From those results Enlund concludes that the first fall in resistance is caused by the transformation of martensite into troostite, and the second by the splitting up of austenite into alpha iron and cementite. Because these curves take into account the increase of resistance which accompanies rise in temperature, the breaks in the curves correspond only with certain maxima reaction velocity; they therefore create a false impression of discontinuous changes during heating. In order to eradicate the effect of the change of resistance due to heating, Enlund tempered his steels for short intervals at gradually rising temperatures, and measured the resistance at room temperature after each tempering. The results obtained by this latter method make it clear that the two reactions which give rise to the anomalies in the continuous curves do not take place suddenly, but are spread over considerable temperature intervals. At the same time, however, the results show that the resistance of the quenched steels does fall in two steps due to the tempering of martensite and austenite respectively.

It is this substantial similarity between the curves taken continuously during heating and the curves determined point by point after periods of tempering, which led Enlund to expect an analagous relation between specific volume measurements made after intermittent heating, and dilatation curves.

But analogy can only be justly and effectively made use of when every possible variable has been accounted for. In this case Enlund has drawn the obvious conclusion, but not the only one. The comparison between resistance measurements and volume measurements is rendered ineffective because a fall of resistance accompanies the change of austenite whatever the resulting constituent, whereas an appreciable increase in specific volume is only brought about by the formation of martensite. Thus the similarity between the electrical resistance results obtained by the two different methods is due to the progressive

progressive tempering of austenite, and the difference between the specific volume and dilatation results is due to the formation of martensite in one case and troostite in the other.

What has been said with regard to the relative tempering rates of austenite and martensite does not apply only to tempering at 240° C. Enlund's results show conclusively that martensite tempers more rapidly and at lower temperatures than austenite. Further, reference to the curves (figure 1) showing the specific volume changes taking place during tempering at different temperatures of a chromium steel containing 1.73 per cent of carbon, will show that at 100° C., 150° C. and 200° C. the specific volume decreases. This decrease is the result of the tempering of a small amount of martensite in the steel. The rise in specific volume observed after tempering at 250° C. indicates the commencement of austenitic tempering. These results are in close agreement with those of Enlund, and with those of Andrew and Honeyman (1). All of them go to show that the rate of tempering of austenite does not, at any temperature, equal or exceed the rate of tempering of martensite.

As a further example of this formation of martensite during cooling, the specific volume measurements made before and after immersion in liquid air may be cited.

The specific volume of the quenched chromium steel containing 1.73 per cent of carbon increased considerably after being immersed for one hour in liquid air. The straight carbon steel with the same carbon content was similarly affected. (see table 12). This rise in specific volume is clearly the result of austenite changing to martensite at these low temperatures, and is further evidence that martensite forms only during progressive cooling. A quenched steel may be considered as one tempering at atmospheric temperature, so that immersion in liquid air is equivalent to cooling down from the tempering temperature.

When the electrode potential (table 11) and the electrical

electrical resistance (table 12) of austenitic steels were measured before and after immersion in liquid air, no considerable change was noted. From this it may be concluded that austenite can change to martensite without any apparent change in resistance. This is in accordance with the view advanced in the electrical resistance section, where it is stated that austenite and martensite have practically the same specific resistance when they contain the same amount of carbon in solution. At atmospheric temperature the electrical resistance is only slightly affected by the change from gamma to alpha iron, so long as no carbon separates in the process.

In connection with this point it may be recalled that the conclusion drawn from the tempering resistance curves with regard to the relative tempering rates of austenite and martensite, was not based on any distinction between the specific resistances of these constituents, but on the different rates at which they changed.

It seems to be fairly well established that there is no sudden change from austenite to martensite during tempering: that there is no progressive accumulation of martensite during tempering at a constant temperature: and that the martensite which results from the decomposition of austenite does so during cooling from the tempering temperature. This means that during progressive cooling austenite transforms to martensite at a much greater rate than martensite breaks down to troostite, otherwise no accumulation of martensite can result. But it has been stated that at any constant temperature austenite transforms much more slowly than martensite. Therefore the mechanism of the transformation of austenite during progressive cooling is conspicuously different from the process of tempering at constant temperature. Although these views will be developed later, it is perhaps not inappropriate to mention some aspects of the matter at this point. The two different ways in which the austenitic

austenitic transformation can take place are considered by the writer to correspond with the Ar' and Ar" of Portevin and his collaborators. The Ar', which is the Ar 1.2.3. change, gives rise to the production of pearlite, sorbite or troostite, depending on the temperature at which it occurs: whereas the Ar", which is the Ar 2.3. change, gives rise to martensite, troosto-martensite or troostite, depending on the range of temperature over which it spreads. The Ar' change proceeds at its maximum velocity in the neighbourhood of 700°C. and thereafter, as the temperature is lowered, it proceeds more slowly. The Ar" change does not occur at all at high temperatures, but takes effect when the transformation tension of the allotropic change is sufficient to overcome the inhibitory effect of the dissolved carbon. Its commencement will therefore be determined by the amount of carbon dissolved, and the range of temperature over which it extends will be determined by the concentration gradient of the dissolved carbon. If cooling is suddenly stopped at some point in the temperature range of Ar", the change will proceed at the rate and in the manner of Ar', unless the temperature is so low as to inhibit Ar' entirely.

There is no discontinuity between the changes which take place during tempering and those which take place during quenching. When austenite, which has been retained by quenching, is tempered, the mechanism of the change is the same as that which gives rise to the formation of troostite or sorbite during cooling. When the austenite is cooled from the tempering temperature, martensite is formed in the same way as during quenching.

The traditional view, that quenching is a means of retaining the state which existed at the quenching temperatures, is very persistently held, and firm adherence to this conception of quenching has been responsible for much confusion in the past. Now the conclusions which have been drawn from the tempering-resistance and the specific volume results show clearly

clearly that martensite is produced only during cooling, and not at the tempering temperature. Cooling therefore produces, in this instance, a constituent which otherwise does not appear. When austenite is retained by quenching, then and then only can it be said that the quenching has retained the state which existed at the quenching temperature. All other constituents and complexes of constituents are the result of the lowering of the changes to temperatures below the normal temperature of the change.

Much of the work which has been performed in order to ascertain the relations of graphite and cementite, has been based on the assumption that rapid cooling or quenching serves to retain the state existing at the quenching temperature. But as has pointed out elsewhere by Mr. M. S. Fisher, such methods have little meaning. The fact that cementite results from rapid cooling and graphite from slow cooling, does not show that cementite is a transient constituent which later decomposes into graphite. All that can justly be concluded from the results obtained by quenching cast iron, is that cementite forms during rapid cooling and graphite during slow cooling, and that the two constituents, graphite and cementite, have the same relation to each other as pearlite and troostite in a steel. They are, in fact, two different products of the same change taking place under different conditions.

Some experiments described by Griffiths (47) may be referred to here as they serve to throw additional light on the changes taking place during cooling, at a rate sufficient to produce the Ar" change. A specimen of a nickel-chromium steel was cooled from 900° C. at such a rate that the Ar' was suppressed, and was quenched from just above the Ar"point. A second specimen was cooled from 900° C. in a similar way to the middle of Ar" and quenched, while a third specimen, after a similar treatment, was quenched just at the end of the point. Examination of

of these three specimens showed that the amount of martensite decreased as the quenching temperature was lowered, whereas the amount of troostite increased. From these results Griffiths concludes that the so-called troosto-martensite is formed during the Ar" and not below it by the tempering of martensite. He also admits the possibliity that the martensite may be a product of some lower change point, although he himself does not incline to this view. These results and conclusions of Griffiths are in entire agreement with the views set forth in the present paper, but in order to bring them into line with what has been said previously a slightly different explanation may be attempted.

During progressive cooling through the temperature interval of the Ar", austenite breaks down progressively to martensite. For reasons which will be fully set out in another section, the rate of cooling through this temperature interval will have little effect on the change of the austenite to martensite. Therefore the austenite-martensite transformation will take effect whether the specimen be quenched from just above the point or slowly cooled through it. But the rate of cooling through the point will have a noticeable effect on the subsequent tempering of martensite to troosto-martensite or to troostite. In fact at the beginning of the change, which in the case of the steel examined occurs at 500° C., the rate of tempering of martensite will be so great as to cause the martensite to break down to troostite as quickly as it is formed. If, however, the steel is quenched from above the point, the martensite will form as before, but the greater cooling velocity will prevent the simultaneous decomposition of this martensite. On the other hand, if the steel is cooled to the middle of the Ar" change before quenching, then some of the martensite first formed will have decomposed as it formed. When the steel is cooled slowly through the entire change and quenched from 340° C., most of the martensite formed during the change will have decom-

decomposed, but as the rate at which this decomposition proceeds is continuously falling off as the temperature drops, some of the martensite last formed will be retained.

Troostite and troosto-martensite can be produced at the Ar" as well as martensite. Which constituent will result depends entirely on the rate of cooling through the point and the temperature at which the point takes effect.

Another outstanding feature of the tempering-resistance curves is the tendency of these curves to become horizontal, long before the resistance characteristic of these steels in the annealed state is reached. The resistance of each steel does in fact eventually attain a value, which, under the circumstances, may be assumed to be characteristic of the tempering temperature. Electrode potential and specific volume measurements, made after tempering for long periods at various temperatures, indicate similar tendencies, and in general all tempering experiments serve to show that at each particular temperature, the process of tempering proceeds towards the annealed state only to a certain extent. It cannot be safely presumed that this state, attained after fairly long tempering, constitutes the ultimate stable state, but it can be stated definitely that further progress towards the ultimate stable state takes place very slowly.

It appears, from consideration of the electrode potential and specific volume results, that this state attained by tempering approaches nearer and nearer to the annealed state, as the temperature of tempering is raised. And further it is evident, from consideration of the resistance of steels fully tempered at 240° C., and of the specific volume of steels tempered at progressively rising temperature, that the tempered state differs from the annealed state by approximately the same amount for all steels. This point is clearly shown in figures 19, 20 and 21, where the resistance of the steels in the quenched

quenched, tempered and annealed states is shown. In these figures the line representing the resistance of the tempered steels is nearly parallel to the line representing the resistance in the annealed state. The specific volume experiments do not point so conclusively to this parallelism, but even with them there is an evident tendency to reach a line parallel to the annealed line.

There is therefore reason to suppose that there is something fundamental about the state attained by tempering at a particular temperature. This state is apparently different for each particular temperature, and is more closely related to the annealed state than to the quenched state. Nothing of the nature of a phase in donformity with the requirements of thermodynamics is here suggested. Whatever there is of stability or singularity in this tempered state, must be attributed to a balance between the forces which strive for the attainment of the true stable state and the viscosity or other inhibitory factors arising from the abnormal conditions existing.

SECTION 4. SUB-SECTION C.

CONTINUOUS MEASUREMENT OF ELECTRICAL RESISTANCE DURING TEMPERING.

MATHEMATICAL CONSIDERATION: 1st METHOD.

MATHEMATICAL EXPRESSION OF THE TEMPERING RESISTANCE CURVES.

First Method.

Consideration of the curves showing the change of resistance during tempering would be greatly simplified if they could be expressed in a mathematical form, but on account of the present disagreement in theories of tempering, it is a matter of some difficulty to find a physical basis for mathematical treatment.

W. Fraenkel and E. Heymann, whose work is referred to in another part of this thesis, assumed that their tempering resistance curves could be expressed according to the formula for the rates of reversible reactions. They found that the curves approximated to the formula $\frac{dc}{dt} = KC^{10}$ and deduced from this that the tempering of martensite is a reaction between iron and carbon to form FeC_{10}. The present writer considers that this interpretation is based on an incorrect conception of the process of tempering.

The tempering of steel appears to be a complex process consisting of a number of simultaneous or concurrent operations such as: the decomposition of martensite and progressive coagulation of its decomposition products accompanied by some diffusion of carbon: and complex decomposition of austenite, either direct or through martensite, into alpha iron and free cementite.

When no appreciable amount of austenite is present, the form of the curve is dependent on the amount of carbon dissolved in the martensite. Steels which contain austenite give a tempering curve which, during the early stages, represents the sum of two reactions proceeding simultaneously, but, when all the martensite originally present has gone, the curve is that of pure austenite. The austenite curves likewise vary with the carbon content of the austenite, but it is difficult to

to differentiate between the effect of increasing amounts of austenite and that of greater quantities of carbon dissolved in the same amount of austenite.

As would be expected, the tempering temperature also affects the tempering curves, and although the present work was carried out at one temperature only, an idea of the temperature effect may be obtained from the paper by Fraenkel and Heymann. This effect appears to operate in more ways than one, for not only does it influence the viscosity of the material and so affect the tempering rate, but at the same time it determines the state to which the tempering steel will tend. It is this state that determines the curves, and not the annealed state, as might be expected.

Although the curves shown extend only to 260 minutes, the tempering was carried much farther, as shown in the tables. In every instance the total time of tempering exceeded 400 minutes, and with the high carbon and alloy steels the time was increased to over 600 minutes. Each curve was continued until there was no further appreciable change in resistance, so that the final values obtained approximate very closely to the limiting values for that tempering temperature. This value is a function of the temperature and is as fundamental as the rate at which it is attained.

From consideration of the foregoing, it is apparent that any equation which is to express the tempering curves must take into account the carbon content, the temperature of tempering, and the amount of tempering that can occur at that temperature.

When these factors are suitably expressed and embodied in an equation, it should be possible to calculate the rate at which tempering proceeds. The curve showing the relation between the composition of the quenched carbon steels and their resistance is a straight line, and therefore the resistance of

of any steel in the series is given by an equation of the form

$$R = Mc + N$$

where R is the resistance of the quenched steel

c is the carbon content

M and N are constants.

After tempering fully at 240°C., the carbon steels give a resistance composition curve of a somewhat indefinite form. As the total time of tempering was not identical in all cases, it might be assumed that the resistances measured after tempering should properly lie on a straight line. Unfortunately the results admit of a number of alternative straight lines, and consequently the equation for the resistance of tempered steels is not so unequivocally determined as that for the quenched steels. During subsequent efforts to accomodate the tempering-resistance curves to a suitable formula, all the possible straight lines for the tempered steels were tried. In the course of this work it became apparent that the assumption of a straight line form for the resistance-composition curve was unlikely to give the best results. Accordingly a smooth curve through the points was taken. This curve was found to conform to an equation of the form

$$\bar{R} = Ac^2 + Bc + K$$

where $\bar{R}$ is the resistance of the tempered steel

c is the carbon content

A. B. and K. are constants.

The two curves, one representing the resistance of the quenched steels, and one representing the resistance of the tempered steels, meet at a point in the neighbourhood of .16 per cent carbon and 14 microhms. If this point is considered as the origin, then K and N the constants in the two equations, become equal. Whether this point has any significance from a physical point of view is immaterial at present. All that need be said is that it is convenient to use it as the origin.

A general survey of the tempering-resistance curves will indicate that the fall in resistance within a specific time interval is proportional to the amount of carbon in the steel, and to the length of time during which tempering has been proceeding. These facts are obviously the determinants of the curves and must form the basis of any equation which is designed, not only to meet the curves, but to convey their significance. Starting from this basis a number of equations in which the above facts were expressed in one form or another was tried, and eventually an equation of the form

$$R_o - R_T = \frac{x}{1 + \frac{b}{T}}$$

was selected as combining comparative simplicity with applicability. In this equation

R_o is the resistance of the steel at zero time at 240 C.

R_T is the resistance of the steel at time T minutes after commencement of tempering.

b is a factor depending on the tempering temperature

x is a factor depending on the carbon content

When T is very large R_T approaches the limiting value for the tempering temperature, that is, it approaches $\bar{R}$, so that

$$x = R - \bar{R}$$

but

$$\bar{R} = Ac^2 + Bc + K$$

and $R_o = Mc + N$

therefore $x = (M - B)c - Ac^2$.

As stated above, the constants K and N are equal.

For the carbon steels A = -2.23. B = 9.56. M = 27.5. and c = the carbon content less .16 per cent, this percentage being the carbon content at which the lines representing the resistance of the quenched and tempered steels respectively, meet. Substituting these values in the above equation x is

is found to be given by

$$x = 17.94\ c + 2.23\ c^2$$

By substituting the appropriate carbon content in this equation, x is given for each steel. All the factors in the general equation for the tempering-resistance curves, with the exception of the constant b, are therefore known. This factor b, which is dependent on the tempering temperature, can only be found by applying the equation to the actual curves. By finding x for each steel and taking values of $R_o - R_T$ and T for a number of points on each curve, the value of b was found to be 1.5 for all carbon steels. Besides b, which depends on the tempering temperature, x is the only factor to be found. As x can be found from the curves of the resistance of the quenched and tempered steels, and as this resistance depends on the carbon content, it follows that the tempering resistance curves are fully determined by the carbon content and the temperature of tempering.

For the nickel steels the formulae are similar.

$$R_q = 29.5\ c + 23.5$$

$$\bar{R} = 5.08c^2 + 12.1c + 23.5$$

Therefore $x = 17.4c + 5.08c^2$.

In this case the curves representing the resistance of the quenched steels and the tempered steels respectively, meet at .2 per cent carbon, so that c is the carbon content less .2 per cent. The constant b has the same value as in the equation for the carbon steels.

When the deduced values of the factors were substituted in the general equation and the equation used to determine the tempering resistance curves, it was found that there was substantial agreement between the curves so determined and the experimental curves. Some of the calculated curves are shown alongside the experimental curves in the figures. All the

the experimental curves in the carbon series, except the curve representing the tempering of the steel containing 1.61 per cent of carbon, are in close agreement with the calculated curves. The reason for the divergence of the experimental curve of the steel containing 1.61 per cent of carbon from the calculated curve, is to be found in the presence in this steel of a considerable amount of austenite. The equation derived applies only to martensitic steels, and any steel which contains a noticeable amount of austenite will give a tempering-resistance curve entirely different from the curve calculated by means of the formula. With the data available it is not possible to derive an equation for austenitic steels by the same method as was used to obtain the equation for the martensitic steels.

In the nickel series the divergence of the determined curves from the calculated curves appears at a much lower percentage of carbon. Thus nickel steels containing .87 and 1.06 per cent of carbon respectively, must contain austenite.

As already stated, the temperature at which the tempering is carried out is an important factor. Unfortunately only one tempering temperature was used in the present investigation. Fraenkel and Heymann, however, give data for other temperatures, and express their results as a percentage of the difference between the resistance as quenched and as annealed. These authors assume that on tempering at any temperature the steel will finally attain the fully annealed state. This is evidently not the case, and it is not possible to deduce the real limiting values from their data. If, however, some approximation to this limiting value is assumed, it is possible by means of their percentage figures to gain a general idea of the variation of the rate of tempering at different temperatures. In figure 60 a series of curves showing the approximate tempering rates at different temperatures of a steel containing 1.19 per cent of

of carbon is plotted.

By applying the general formula for the tempering of martensitic steels to these curves, it was found that the constant b increased rapidly with decrease in the tempering temperature. No accurate value of b for these lower temperatures could be obtained, as it was found that the variations of this constant with time were considerable. These variations, which were inconsiderable at high temperatures such as 240°C. cannot be overlooked when b is very large.

SECTION 4. SUB-SECTION D.

CONTINUOUS MEASUREMENT OF ELECTRICAL RESISTANCE DURING TEMPERING.

MATHEMATICAL CONSIDERATION: 2nd METHOD.

MATHEMATICAL EXPRESSION. SECOND METHOD.

Although the formulae devised by the foregoing method of mathematical analysis express the curves with considerable accuracy, the precise significance of the factors in the equations is not apparent. The formulae are comparatively complex and do not serve to show the essentially simple nature of the tempering process.

As already pointed out, the tempering of steel must be considered to consist of a number of closely related reactions taking place simultaneously and consecutively. Each of these reactions is in itself simple but the cumulative effect, as observed by measuring the change in electrical resistance, is not immediately reducible to simple terms. If it were possible to isolate these reactions and watch the progress of each when separated from the other reactions which neutralise and reinforce it, the simplicity of the process would be more evident. Because of the mutual dependence of the reactions, it is not possible to measure their progress separately and recourse must be had to some method of resolving the resultant curve into suitable components. Such a method has proved satisfactory in analysing the resistance-tempering curves. This method will now be described.

The martensitic and austenitic steels give rise to conspicuously different types of curves and must be considered separately. Exact knowledge of the behaviour of martensite is necessary before attempting the analysis of the curves for steels containing both martensite and austenite: it is therefore convenient to consider first the curves for the martensitic steels. The process of tempering in these steels may be divided into three distinct positions. There is in the first place the separation of carbon from solution accompanied by a fall in resistance proportional to the amount separating. As the

the carbon separates it is not removed from the system but remains as free carbide. The formation of this free carbide by the separating carbon results in an increase in resistance proportional to the amount separating. This constitutes the second effect of tempering.

These two effects are mutually dependent, but as they tend to neutralise each other they give rise to a resistance curve which does not show the simple nature of either. Therefore for reasons which will be emphasised presently, the work of interpreting the tempering resistance curves will be greatly facilitated by separating the effects of these two reactions.

The state attained at the conclusion of this primary separation of carbon is not permanent, but the subsequent change takes place more slowly and can be conveniently separated from the first rapid change. This transient state attained at the conclusion of the period of rapid change may be called the primary tempered state. Thereafter a gradual fall of resistance occurs, due to the slow change of the steel from this primary tempered state towards some ultimate stable state - probably resembling the annealed state.

Although for the purpose on hand it is sufficient to recognise the existence of this change without considering its mechanism, a short account of the internal changes which give rise to this resistance change may be attempted. It seems that the greater the vicosity of the material in which the separation takes place - that is the lower the tempering temperature the greater will be the difference between the primary tempered state and the ultimate stable state. Briefly, this difference will be a difference in completeness of separation associated with a difference in size of the separated particles. Subsequently, slow diffusion between the imperfectly separated constituents will proceed accompanied by an increase in the size of the separated particles and by a steady fall in the resistance.

These three simple processes combine to give the observed tempering curves, and the method by which these curves may be resolved into appropriate components will now be described.

In the first place, the point at which the first rapid change is completed must be decided. As the two reactions overlap and lose their identity imperceptibly, the selection of this point is purely arbitrary. Thus any point between twenty and thirty minutes would be suitable. For the present purpose twenty six minutes was chosen as the junction between the different rates of change.

With the assistance of the known values for the different carbon steels as quenched and after tempering for twenty six minutes, the first rapid portion of the curve may easily be divided into the two components representing respectively, fall of resistance due to carbon separating from solution, and rise of resistance due to the formation of carbide areas by this carbon and the requisite proportion of iron. This was accomplished in the following way:-

The values for the specific resistance at 240° C. of the quenched carbon steels, that is the resistance at zero time on the tempering - resistance curves, were plotted and the best straight line through the points was drawn. This straight line represents the effect of carbon in solution on the resistance of iron at 240° C. Similarly, the best straight line through the values for quenched carbon steels tempered for twenty six minutes at 240° C. represents the effect of carbon out of solution on the resistance of iron at 240° C. These two straight lines meet in the neighbourhood of the point .24 per cent carbon and 30.5 microhms. Evidently when less than .24 per cent of carbon is present, its effect on the resistance is less when in solution than when out of solution, no doubt because of its mode of distribution in the metal.

In figure 40 these two straight lines are shown together

together with the best line through the quenched values for carbon steels measured in the cold, and the values measured under the same conditions at the conclusion of tempering.

From these lines it may be deduced that the specific resistance of a steel containing 1.73 per cent of carbon in solution would fall from 73 microhms to 30 microhms in twenty six minutes , if the carbon as it separated was removed from the system. But the separating carbon is not removed from the system, and its effect in raising the resistance is represented by the line through the values for steels tempered twenty six minutes at 240°C. Thus, the specific resistance of the steel containing 1.73 per cent of carbon would be raised from 30 microhms to 47 microhms by the presence of 1.73 per cent of carbon out of solution. The total effect of the tempering as measured experimentally is therefore a drop of 26 microhms, made up of a fall of 43 microhms and a rise of 17 microhms.

To resolve the experimentally determined tempering curves into two components it is therefore necessary to multiply the observed drops by $\frac{43}{26}$ to obtain the fall of resistance due to separation of carbon, and by $\frac{17}{26}$ to obtain the rise of resistance due to the free carbide formed. If the rate at which carbon comes out of solution is proportional to the amount of carbon in solution, the components representing the fall of resistance due to separation of carbon from the various carbon steels should all conform to the same general curve. On resolving the tempering-resistance curves for the martensitic carbon steels into two components in the manner described, it was found that these components did conform to a general curve.

In the carbon series the steel containing 1.19 per cent of carbon is the highest in the series which consists of martensite only: therefore the general curve obtained as above cannot be used to show the separation of carbon from steels containing more than this amount. The principal object of analysing these

these curves was to find out the exact relation of carbon content to the tempering of martensite, so that the curves for steels containing both martensite and austenite might in turn be resolved into their components. Therefore the general curve for fall of rssistance due to separation of carbon, must be continued back from 1.19 per cent to 1.73 per cent of carbon. If the rate of separation of carbon depends on the amount of carbon in solution, the resistance of the steels in the carbon series will be represented at any time between zero time and twenty six minutes, by a straight line through the pivot point at 30 microhms and 2.4 per cent carbon. It can therefore be assumed that after a certain interval of time, the resistance of the steels in the series will have altered, so that the line BO figure 40 represents the best line through them. There is ample justification for this assumption, for not only do the present results point to such an interpretation of the tempering process, but the figures given by Campbell (48) point to a similar conclusion. (figure 41).

If then the resistance of the carbon steels after a certain interval of time can be represented by the line BO, the form of the general curve for the fall of resistance due to separation of carbon can easily be ascertained. By stepping down the line as shown in figure 40. the drops of resistance in successive equal intervals of time may be determined. Such a curve should lend itself to easy mathematical expression, but for the present purpose the geometrical method was found convenient. A number of curves with a progressively increasing scale for carbon and resistance were used in order that the readings towards the end of the curve might be obtained accurately. The vertical lines between AB and BO represent the successive drops in specific resistance in successive equal intervals of time. The horizontal lines between AO and BO represent the successive drops in the amount of carbon in solution,but, as the fall in resistance

resistance is proportional to the fall in carbon, the curve for the rate at which carbon separates from solution can be more easily obtained by multiplying the resistance drops by the appropriate factor. As already stated, when the experimentally determined curves were resolved into their components, the components, representing the fall of resistance due to separation of carbon, conformed to a general curve. This general curve plotted on a suitable axis could be superimposed on a portion of the theoretical curve derived as described above. That is to say, the general curve derived theoretically and the general curve derived from the experimental results could be made to coincide over the interval representing the tempering of a 1.19 per cent carbon steel. The time intervals on the curve derived from the experimental tempering-resistance curves being known, the time intervals on the theoretically determined were obtained. Plotted on the proper time axis, this theoretically derived curve represents the rate at which the resistance falls at 240° C. due to carbon coming out of solution. This curve is shown in figure 42.

In a similar manner the general curve representing the rise in resistance due to the effect of the separated carbide was obtained, and is shown in figure 43.

From the curve representing the fall of resistance due to separation of carbon from solution, the rate at which the carbon comes out of solution was determined by multiplying the drops in resistance per minute by the factor expressing the ratio between carbon in solution and resistance. This factor was of course obtained from the slope of the line 40 in figure 40 from which it was found to be .03334. This curve, representing the rate at which carbon comes out of solution at 240° C. is shown in figure 44.

By the same method the corresponding curves for the nickel, chromium, and nickel-chromium steels were obtained.

obtained. The curves showing rate of fall of resistance due to separation of carbon, rate of rise of resistance due to effect of separated carbide, and rate at which carbon comes out of solution, are shown in figures 42, 43 and 44 respectively.

By means of the curves plotted in figures 42,43 and 44, it is possible to construct the tempering-resistance curve for any steel within the limits of composition of the steels examined. As a check on the work, the curves derived in this way were compared with the tempering-resistance curves actually determined. In figures 45, 46, 47, 48, this comparison is shown for the carbon, nickel, chromium, and nickel-chromium steels. The full lines represent the derived curves. The points are those experimentally determined.

Except for a few points at the beginning of the curve, the agreement is quite satisfactory. That some of the first points fall a little below the theoretical curve is probably a result of the method of obtaining the theoretical curve. The line AO in figure 40 represents the best line through the first values on the curve. These first values were not obtained with the remainder of the curve. They were calculated from the known resistance at 15°C. and the temperature coefficient of resistance. It is not surprising that there is, in some instances, a noticeable discrepancy between these values and the first values measured in the tempering bath.

The tempering-resistance curve for the carbon steel containing 1.61 per cent of carbon as determined by means of the three general curves described, is entirely different from the curve obtained experimentally. This steel is only distinguished from the remainder of those in the carbon series by the presence in it of a considerable amount of austenite. It may therefore be presumed that the difference between the experimental curve and the curve obtained by the general method is due to the presence of this constituent. In figure 46 these two curves are shown.

shown.

As this steel, when quenched, is not entirely austenitic the experimental curve does not represent the tempering curve of pure austenite. It represents in fact, the tempering curve of a mixture of austenite and martensite in unknown proportions. The other curve - that which was obtained by the general method - does represent the tempering curve of martensite in a steel containing 1.61 per cent of carbon. If it can be discovered how much this latter curve contributes to the former, then the curve showing the tempering of austenite in a steel containing 1.61 per cent of carbon will be obtained. In other words, the experimental curve for this steel is the resultant of two curves, one of which is known. The other is to be found.

If the steel consisted entirely of martwnsite, the specific resistance would fall from 70 microhms to 49.4 microhms in the first twelve minutes after the commencement of tempering. According to the curve obtained experimentally, the resistance only falls from 70 microhms to 53.5 microhms in this interval of time. The greater portion of this fall of resistance taking place in the first twelve minutes, is due to the tempering of the martensite which tempers very rapidly. But it is not all due to this. Some austenite will also temper and contribute to the total change in resistance.

The only way in which this change due to austenite may be discovered is by extrapolation. In the case of the steel containing 1.61 per cent of carbon, the extrapolation is difficult because of the large change taking place in the first twelve minutes. In the case of the special steel it is much easier and more accurate.

However, if the curve is extrapolated back from twelve minutes to zero time it meets the resistance axis at 57.5 microhms. The change of resistance arising from the tempering of austenite in the first twelve minutes is therefore 4 microhms. The

The change in resistance due to the tempering of martensite is therefore 12.5 microhms. If the steel had been entirely martensitic the change would have been 20.6 microhms. The change in resistance due to the tempering of martensite in this steel is $\frac{12.5}{20.6}$ of the total martensitic change for a steel of this composition.

By taking the drops in resistance in intervals of one minute on the martensitic curve, and multiplying each drop by $\frac{12.5}{20.6}$, the martensitic component of the experimental curve is obtained. By subtracting the drops per minute on the curve representing the martensitic component from the actual drops on the experimental curve, the austenitic component is obtained. This austenitic component bears a similar relation to the curve which would be obtained if the steel consisted entirely of austenite, as the martensitic component does to the pure martensitic curve. Thus, to obtain the curve which would represent the fall of resistance if the steel were entirely austenitic, the drops per minute on the austenitic component must be multiplied by $\frac{20.6}{8.1}$. When this was done the third curve in figure 45 was obtained. This curve (No.3 figure 45) represents the rate at which the resistance of a steel containing 1.61 per cent of carbon changes during tempering at 240° C.

In the same way as the curves for the tempering of pure martensite were capable of resolution into three components, so the austenitic curves may be resolved into three components. The relation of these components is not exactly the same in the austenitic steels. As the tempering of austenite takes place much more slowly, the rise of resistance due to the formation of cementite and the slow fall due to the subsequent change from the primary-tempered state to the ultimate stable state, cannot be separated. Both reactions proceed simultaneously. The third component, representing the fall of resistance due to the separation of carbon from solution in gamma iron, may be determined

determined as for martensite.

By the same method the austenitic resistance-tempering curves for the other austenitic steels were determined. The martensitic curve, experimental curve, and austenitic curve for each of these steels is shown in figures 47 to 50. The components, representing the fall of resistance which accompanies the separation of carbon from solution in austenite, were also calculated and are shown in figure 51.

As the fall of resistance bears a constant ratio to the carbon coming out of solution, the rate at which the carbon separates may be calculated for each steel. When the curve representing the rate of separation of carbon was plotted on a suitable time axis, it was found to coincide with the general curve for the separation of carbon from martensitic carbon steels. Thus, the separation of carbon from solution in gamma iron proceeds in the same manner as the separation of carbon from solution in alpha iron, but at a lesser rate. In figures 52 to 55 the comparison between these curves and the general martensitic curve is shown. From these figures it will be seen that the austenite in a nickel steel containing 1.06 per cent of carbon tempers 18 times more slowly than a martensitic carbon steel: that the austenite in a chrome steel containing 1.73 per cent of carbon tempers 66 times more slowly: that in a nickel chrome steel containing 1.28 per cent of carbon it tempers 80 times more slowly: and in a carbon steel containing 1.61 per cent of carbon it tempers 10 times more slowly. Incidentally, it may be observed that the austenite in the nickel-chrome steel containing 1.5 percent of carbon does not temper at all at 140° C. All these curves representing the rate at which carbon separates from austenite are shown together in figure 56.

The curves in Figure 60 show the rate at which a martensitic steel containing 1.19 per cent of carbon tempers at various temperatures. The curves were calculated from the figures of Fraenkel and Heymann (49). These authors give a table, in which the amount of change which takes place during definite time intervals at the different temperatures is expressed as a percentage of the total change. By applying these percentage figures to the tempering-resistance curve of the carbon steel containing 1.19 per cent of carbon, the curves in Figure 60 were obtained. In Figure 61 the variation in the rate of tempering with tempering temperature is shown. This figure has only a comparative meaning, and in making it the rate of tempering at 240° C. was taken as 1.

SECTION 5.

DILATATION EXPERIMENTS.

DILATATION EXPERIMENTS.

In this section a few dilatation experiments are described. The curves obtained are examined in their relation to the iron carbon diagram and to current views on the low temperature changes. As there are two distinct sets of curves, those for plain carbon steels and those for nickel-chromium-molybdenum steels, consideration of the work falls naturally into two sections. In the first section the heating curves of the carbon steels will be considered from the point of view of the diagram. In the second section the cooling curves of the molybdenum steels will be considered from the point of view of low temperature changes.

DESCRIPTION OF EXPERIMENTAL METHOD.

The dilatometer used in the present work was similar to that described by Andrew (50). The quartz rods, however, were made much longer in order to have the water reservoir further from the furnace, and the diaphram of the reservoir, against which the quartz rod pressed, was made of a larger diameter to attain greater sensitivity.

Round specimens 2 inches long and $\frac{5}{8}$ inch in diameter were used. The centre of the specimen was bored out to a diameter of $\frac{1}{4}$ inch, and a slot in the side of this hole provided a suitable place for the couple.

Heating was carried out in a nichrome wound electric furnace. No vacuum or other artificial atmosphere could be used, so heating was carried out in air. Temperatures were measured by means of a potentiometer and galvanometer. Expansion was measured by means of scale, graduated in metres and fractions, fixed to the narrow bore tube leading from the reservoir. The readings on this scale are the ordinates in the dilatations curves shown. By means of a standard copper specimen the absolute value of the scale readings was obtained. It

It was found that 1 centimetre on the scale equalled an actual expansion of .0024 inches.

DILATATION CURVES OF CARBON STEELS.

One heating and one cooling curve was taken of each of the steels in the carbon series. The heating in all cases was rapid.

The main features of the curves are similar to those usually obtained with a direct measuring dilatometer. They are strictly in accordance with what would be expected from the equilibrium diagram, and a closer analysis will give to this statement a quantitative meaning.

If the heating curve for the steel containing .22 per cent of carbon (figure 62) be considered, it will be seen that the change point is spread over a considerable range of temperature. Now during the change the steel would still be expanding due to rising temperature. At the same time it would be contracting due to the allotropic change. Therefore the true magnitude of the change will not be evident immediately.

To obtain the real value of the contraction which has occurred, it is necessary to make allowance for the normal expansion of the steel within this range of temperature. This can most readily be done by extrapolating the line representing the subsequent expansion to a point below the commencement of the change, and measuring the vertical distance from the commencement of the change to this extrapolated line. This is shown in figure 62 from which it will be seen that the total contraction at 700° is 30 centimetres on the scale used. Now this procedure is only correct if the temperature coefficient of expansion is the same before the change as after it. If there is any appreciable difference between the coefficients, it is not accurate to assume that the steel expanded during the change at a rate represented by the temperature coefficient

coefficient of expansion after the change. For strict accuracy it would be necessary to vary the coefficient in proportion to the amount of the change which has taken place. However, as the two coefficients are approximately the same in this particular case, there is no serious error introduced by proceeding as described.

If the curves from the steels containing .48, .71, and .9 per cent of carbon respectively are treated in the same way, it appears that, as the percentage of carbon is increased, there is a gradual decrease in the amount of contraction taking place at the change point. These curves are shown in figures 63, 64 and 65, and it will be seen that not only does the contraction get smaller, but the range of temperature within which it occurs decreases also. This of course demands no other explanation than is provided by the equilibrium diagram. Solution of carbon, as is well known, expands the austenitic lattice. If this solution takes place at about the same time as the allotropic change, the expansion due to solution will mask the contraction due to the allotropic change. This is what happens in hypo-eutectoid steels with the result that the change point, as measured dilatometrically, decreases in magnitude as the carbon content increases.

With hyper-eutectoid steels only that carbon which is part of the pearlite will be dissolved at the change point. The carbon in excess of .9 per cent will be dissolved progressively with rising temperature. Dilatation curves for these steels should therefore contract at the change point to the same extent as a .9 per cent steel, and should expand thereafter due to solution of more carbon. Ultimately, when all carbon has been dissolved they should continue to expand at a rate determined by the coefficient of expansion characteristic of the steel. Reference to figures 66, 67 and 68 will show that the contraction at the change point is similar to that for a pearlite steel,

steel, and that the subsequent expansion takes place at a substantially greater rate.

These are the most obvious features of the dilatation curves for carbon steels. The whole phenomena observed is easily explained by the contraction accompanying the allotropic change and the expansion accompanying the solution of carbon.

Returning to the dilatation curve of the steel containing .22 per cent of carbon, it will be seen that the change point takes place in two definite steps. This is rather peculiar and is deserving of further consideration. At 700°C. the pearlite portion of the steel will change rapidly, but the amount of the pearlite is, in this case, insufficient to delay the rise in temperature of the steel as a whole. Therefore the normal expansion due to rising temperature becomes sufficient to balance the contraction, and the break in the change point occurs. Subsequently, the contraction due to the allotropic change in the ferrite becomes so great as to exceed the normal expansion and so gives rise to a total contraction. Briefly that is what happens. But it must be remembered that there is no actual break in the continuity of the reaction. The commencement of the contraction due to the allotropic change in the ferrite follows immediately on the pearlite change, but the rate at which the allotropic change proceeds becomes greater as the temperature is raised.

The rate at which the allotropic change proceeds in this particular steel was obtained as follows.

As already stated, the total contraction which takes place at the change point is reduced in proportion to the carbon content of the steel. It is therefore possible to calculate what would be the total contraction if the steel were composed entirely of ferrite. Knowing this, and the contraction which takes place when the steel is entirely pearlitic, the contraction of any steel may be divided into two portions - that due to the pearlitic

pearlitic change, and that due to the subsequent change of the ferrite. For example, a steel which contains .22 per cent of of carbon contains 24.5 per cent of pearlite and 75.5 per cent of ferrite. In such a steel the total contraction at the change point will be 24.5 per cent of the total pearlitic contraction and 75.5 per cent of the total allotropic contraction. The pearlitic contraction is, of course, the allotropic contraction less the expansion due to the solution of carbon, but for the present purpose it is convenient to distinguish the pearlitic and allotropic contractions in the above manner. This distinction is convenient because the pearlitic change takes place first and at constant temperature, whereas the allotropic change in the ferrite takes place subsequently and progressively.

As the change proceeds the temperature coefficient of expansion alters continuously. This alteration takes place because the coefficient of expansion of gamma iron is different from the coefficient of expansion of alpha iron, and also because the coefficient of expansion of austenite depends on the amount of carbon dissolved in it.

The changes which take place during the heating of a steel containing .22 per cent of carbon, and which give rise to the type of curve shown in figure 62, will now be considered in detail.

When this steel reaches 700° C. the pearlite portion changes to austenite with .9 per cent of carbon in solution. As the steel contains 24.5 per cent of pearlite it should immediately contract 24.5 per cent of the total pearlitic contraction. Under the particular conditions of heating employed in the present work, the absorption of heat accompanying the change of this amount of pearlite is evidently insufficient to delay the rise in temperature of the whole steel. Consequently the curve in figure 62 does not show the pearlitic change taking place at constant temperature. It may be assumed that the

the change takes place very rapidly and is completed before the temperature of the whole has appreciably risen. Until this change occurs, expansion will proceed in accordance with the coefficient of expansion characteristic of this steel. Immediately the pearlitic change commences, the coefficient of expansion alters, and for a brief instant the steel will expand at a rate determined partly by the coefficient of expansion of pure ferrite and partly by that of austenite containing .9 per cent of carbon. Thereafter the coefficient ~~of reisitance~~ will change progressively as the carbon diffuses away from the first formed austenitic areas and as the ferrite changes to gamma iron. As these changes are relatively very small it is quite accurate to assume that they take place in definite small steps.

Thus, by assuming that for a few degrees after the pearlitic change had completed itself, the steel expanded according to the composite coefficient of expansion of ferrite and pearlitic austenite, it is possible to ascertain the magnitude of the expansion which would have taken place had no change in the ferrite areas occurred. Then knowing the expansion which should have taken place, and the contraction which has actually taken place, it is possible to calculate the contraction caused by the ferrite which changed during this short temperature interval. From the magnitude of this contraction the actual amount of ferrite which changed in this interval can be found.

By proceeding in this way the amount of ferrite which changes during a number of successive small temperature intervals can be found, and from this a curve showing the rate at which ferrite changes at different temperatures can be plotted. Such a curve computed from the dilatation curve of the steel containing .22 per cent of carbon is shown in figure 70.

Working backwards from this curve, the expansion due to rise in temperature and the contraction due to the change of the ferrite can be computed for small temperature intervals. In

In figure 69 curve BD represents the expansion due to the coefficient of expansion, and curve EC represents the contraction due to the allotropic change in the ferrite. Both of these curves were calculated from figure 70. The sudden contraction DE in figure 69 represents the change due to the transformation of pearlite to austenite at constant temperature.

From BD, DE and EC the curve DF was obtained. This curve, which is the resultant of these already described, represents the change point actually observed. The points alongside DF (figure 69) are those of the original curve for the steel containing .22 per cent of carbon. There is a slight discrepancy between the actual curve and the calculated curve, and is no doubt due to the error introduced by considering the change as taking place in a series of steps. For the purpose on hand, however, the agreement is quite satisfactory.

It is thus possible to give a quantitative meaning to the dilatation curve of this steel containing .22 per cent of carbon, and to show that the iron changes at an increasing rate as the temperature is raised. Rate is used here as meaning the amount which changes within a certain temperature interval, and has no direct connection with time, although the time factor is introduced in the rate of heating through this particular temperature interval.

The fact that the iron undergoes the allotropic change more rapidly as the temperature rises, is worthy of comment. There can be little doubt that this is so because the carbon can diffuse more rapidly at these temperatures. The allotropic change is evidently brought about at these temperatures by the diffusion of the carbon into the unchanged iron and so causing it to change, probably by distending the lattice to a degree equal to the distention normally produced by heating to 900° C. As the temperature rises, the amount of carbon necessary to effect this critical expansion will decrease, for the lattice

lattice will be approaching the critical expansion due to the normal temperature effect. This, together with the fact that the carbon diffuses with increasing facility as the temperature rises, will suffice to account for the increased rate of the allotropic change.

Proceeding to the other hypo-eutectoid steels of the carbon series, it is found that much of the foregoing does not apply. In the steel containing .48 per cent of carbon there is 54 per cent of pearlite. The change at 700° C. should be 54 per cent of the total pearlitic change. Thereafter, the ferrite should change in the same manner and at the same rate as in the steel containing .22 per cent of carbon. On this assumption, it is possible to draw a theoretical curve for the steel containing .48 per cent of carbon. Such a curve is shown in red in figure 71. The points show the actual curve obtained experimentally. The discrepancy between the actual curve and the curve based on the foregoing assumptions, is sufficient to show that either the assumptions are incorrect, or there are other factors operating in this steel. If the conditions existing at the change point in this steel are examined, a number of points of difference between it and the steel of lower carbon content are revealed. In the first place, the amount of pearlite present will cause a much larger absorption of heat, and consequently a longer halt in the neighbourhood of 700° C. This longer halt will enable a greater proportion of the allotropic change to take place before a further temperature rise occurs. Further, the larger amount of carbon diffusing from a greater number of points will tend to speed the change, with the result that the whole is completed in a much shorter temperature interval than would be expected from analogy with the steel containing .22 per cent of carbon.

When the theoretical curve for the steel containing .71 per cent of carbon is compared with the actual curve (figure 72)

72) a similar discrepancy is observed. There is therefore no doubt that the reasoning applied to the curve for the lowest carbon steel is not capable of expansion to meet the curves for the steels of higher carbon content. The shape of the curve shown in figure 69 is typical for steels of low carbon content, and there seems no reason to doubt that the analysis of this curve would be applicable between certain limits of composition. When the carbon passes a certain concentration, the halt period becomes so great that the gradual curve of the volume change is obscured.

From consideration of the variation in the magnitude of the contraction with increase in the carbon content, it appears that one per cent of carbon going into solution expands the lattice by an amount equal to 70 centimetres on the scale. From consideration of the variation of the temperature coefficient of expansion of the austenite with rise in the carbon content, it is evident that 1 per cent of carbon going into solution increases the coefficient of expansion by 10 centimetres per 20 millivolts increase in temperature. If then the rate, in terms of temperature, at which carbon dissolves in hyper-eutectoid steels subsequent to the pearlitic change could be discovered, it would be possible to calculate theoretical curves for the expansion of hyper-eutectoid steels above the change point.

It is evident from the shape commonly given to the S.E. line in the iron carbon diagram, that the rate in terms of temperature, at which solution and diffusion of carbon proceed, increases as the temperature rises. Now the curve already obtained (figure 70) for the increase in the rate of diffusion with rising temperature, although it applies to a limited temperature interval only, might not be entirely unconnected with the general rate of solution and diffusion of carbon above 700° C. Therefore this curve was converted to a suitable axis and plotted in figure 73. By assuming this curve to represent the amount

amount of carbon going into solution at temperatures between 700°C. and 1100°C., and knowing the expanding effect of this carbon on the austenitic lattice and on the temperature coefficient of expansion, it should then be possible to calculate the curves for the hyper-eutectoid steels. This calculation was performed for the steels containing 1.19, 1.61 and 1.73 per cent of carbon. The theoretical curves so derived are shown in figures 74, 75 and 76. The crosses alongside are points on the heating curves actually determined. Although the agreement is fairly good there is an obvious bending away of the actual curve at high temperature. It is very probable that this bending off is due to a slight irreversible contraction of the steel due to the high temperature and moderate pressure existing in this region. In favour of this view it may be stated that there is no appearance of a similar convexity in the cooling curves; that the steel at the conclusion of the cooling is not of the same length as at commencement of heating; and that this tendency for cylindrical metal specimen to become shortened in length by heating, is a commonly observed phenomena. There is therefore evidence to show that the bending off at high temperatures is not due to a reversible constitutional change, but is in effect due to a natural contraction of the specimen brought about either by high temperature alone or by temperature and pressure conjointly.

The circles on the curves 74, 75 and 76 are the actual points on the cooling curve raised a few centimetres to compensate for this contraction. These points agree more closely with the theoretical curve. It may be surmised that this difference in the heating and cooling curves is due to the fact that there is less pressure on the metal during cooling. From the shape of these two curves it ~~will~~ is evident that whatever is the cause of the bending off of the heating curve, there is no similar change taking place in the cooling curve.

Although the agreement between the theoretical and the actual curves is not perfect, it is nevertheless sufficiently close to warrant some faith in the accuracy of the original assumptions.

DILATATION CURVES OF NICKEL-CHROME-MOLYBDENUM STEELS.

The dilatation curves of the nickel-chromium-molybdenum steels are shown in Figures 77 to 86. The analysis of the steel is given in Table 2. Molybdenum and chromium are practically constant in all the steels: the carbon is about .4 per cent in numbers 1, 2 and 3 and about .8 per cent in numbers 4, 5 and 6. The whole series is therefore divided into two series of three steels of similar carbon content. In each of these series of similar carbon content, the nickel varies from 2 to 5 per cent.

Figure 77 shows the effect of initial temperature on the cooling curves of steel No.1 (.37 per cent carbon and 2.02 per cent nickel). Raising the initial temperature lowers the temperature at which the change on cooling commences.

By raising the initial temperature from 770° C. to 1075°C., the temperature of commencement of the change is lowered progressively from 440° C. to 370° C. Whatever the initial temperature, the change on cooling is complete about 320° - 330° C. Evidently, when the beginning of the change is retarded by the effect of higher initial temperature, the change once commenced proceeds more rapidly. This suggests that by raising the initial temperature a more homogeneous solution is formed, and that the temperature at which the change on cooling commences is thereby lowered.

When heterogeneous austenite is cooling, the change will obviously start in those areas in which the concentration of the solute is least. When a more uniform distribution of the solute has been produced by heating to high temperatures or by prolonged soaking, the change will commence less readily, because the areas in which the concentration of the solute is least, will now

now contain relatively more in solution than when the austenite was heterogeneous. At the same time the more uniform austenite will transform within a shorter temperature interval. In heterogeneous austenite the concentration gradient of the solute will tend to prolong the change.

In Figure 78 the effect on molybdenum steel No. 2 of raising the initial temperature is shown. In this case raising the initial temperature from 800°C. to 1035°C. produces a discontinuous lowering of the change point on cooling. This discontinuous lowering from 430°C. to 230°C. would appear to require some other explanation than that suggested for the progressive lowering of the change point of steel No. 1. It does not appear necessary, however, to go outside of the commonly accepted views on heating and cooling for this explanation.

The greater uniformity produced by heating to high temperatures is sufficient to explain progressive lowering within certain limits. To this must be added the effect of the faster cooling, which results when the steel is cooled in the same medium from higher temperatures.

As it is only the commencement of the change which need be considered - once it has started the heat evolved will serve to delay the cooling and facilitate the progress of the change - there appears no reason why it should not be suddenly lowered by raising the initial temperature. It must be remembered that steel is an extremely viscous material, and it is not difficult to conceive of small variations in the distribution of the solute and in the rate of cooling, having a pronounced influence on the position of the change point on cooling.

Figure 79 is very similar to the one just considered. It shows the discontinuous lowering of the change point by raising the initial temperature.

Figure 80 shows the effect of soaking for various times at the initial temperature. The effect is apparently very slight. Soaking for 4 hours at 785°C. (Curve 3) lowers the

the change by the same amount as does raising the initial temperature to 1000°C.

In Figure 81 the effect of prolonged soaking (17 hours) at 730°C. is shown. By this treatment the change point in steel No, 5 is lowered to 400°C., whereas on heating to 810°C. and cooling the change commences at 435°C.

In Figure 82 curves 1 and 2 show the effect on the change point in steel No. 6 of varying the rate of cooling. Curve No. 1 was obtained by cooling with the furnace current on, and with all the external resistance in circuit. Curve No. 2 was obtained by cooling in the furnace in the manner adopted for all the cooling curves, that is, with the heating current shut off. The effect of increasing the rate of cooling in producing a discontinuous lowering of the change point, is the same as the effect of raising the initial temperature.

The remainder of the dilatation curves for these molybdenum steels are shown in Figures 83 to 86. They were all taken with specimens which had been oil quenched from 1000°C. In no case is there any evidence of the expansion caused by the formation of martensite from austenite. The curves all show a gradual bending off, caused by the tempering of the martensite which was produced by quenching.

SECTION 6.

GENERAL CONCLUSIONS.

GENERAL CONCLUSIONS.

Of the results obtained during the present work, those which appertain to the relations between austenite, martensite, and to a lesser extent, troostite, are probably the most significant. They will therefore be discussed first.

SUB-SECTION A. THE NATURE OF MARTENSITE.

Many theories which have been previously advanced to account for the constituent martensite, may now be disregarded. So much of the work done in recent years, and particularly the application of the X-ray spectroscope, has gone to show that martensite is a solid solution of carbon in alpha iron, that this view has come to be held almost universally. But this description is incomplete. Martensite is also a microscopically distinguishable constituent, which, in certain hyper-eutectoid steels, is partly composed of a solid solution of carbon in gamma iron. Sometimes the word is used in a manner that gives rise to confusion as to which meaning is to be read. In the present discussion the word martensite will be used to refer only to a solid solution of carbon in alpha iron, and where this solution is mixed with a solution of carbon in gamma iron, the condition will be defined as partly austenitic and partly martensitic.

Perhaps the term, solid solution, as applied to martensite, is not strictly accurate, but in the present state of knowledge, the term serves to convey a sufficiently appropriate idea of the nature of the constituent. Later, when the nature of solid solutions, inter-metallic compounds, and other constituents of metallic complexes has been more thoroughly investigated by suitable experimental means, facts may be revealed which will warrant a more definite description of martensite, and maybe a distinction between it and other solid solutions. But at present a more detailed definition than the above cannot really be deduced

deduced from known facts, though speculation on the arrangement of the carbon atoms or carbide molecules within the alpha lattice, may proceed unchecked.

Whatever the outcome of future investigations may be, it must at present be admitted that in martensite the iron is in the alpha state and the carbon is retained within it. In short, the component elements have become merged in a single constituent, in such a manner that the identity of each is lost, and new properties are conferred on the resultant single constituent.

Martensite is not the only solid solution of carbon in alpha iron. The ferrite found in annealed and tempered steels, either free or as a constituent of pearlite, sorbite or troostite is also a solid solution of carbon in alpha iron, in which the carbon is present in small and varying quantities.

No doubt the solid solutions of carbon in alpha iron form a continuous series with respect to carbon content: but there is a definite break in the series with respect to mode of origin.

The solid solution found in annealed steels and in steels tempered in the vicinity of the change point, may result either from the break down of austenite, or by solution of carbon in pure alpha iron at a suitable temperature. The solid solution found in quenched steels, and known as martensite, can only be formed from the breakdown of austenite, and never directly by solution of carbon in alpha iron. Martensite is therefore an irreversible solid solution in which the carbon is retained in enforced solution.

Austenite, from which martensite forms, is admittedly a solid solution of carbon in gamma iron. This view of the nature of austenite is explicit in the iron-carbon diagram, and subsequent research has served only to confirm it.

If martensite is a solid solution of carbon in alpha iron, resulting only from the breakdown of a solid solution of carbon in gamma iron, it follows inevitably that formation of martensite

Martensite occurs when the gamma iron of austenite changes to alpha iron, unaccompanied by separation of carbon. Dejean (51) and Scott (52) identify the so called Ar", which gives rise to the formation of martensite, with the allotropic change Ar_3. other workers imply that the facts may be otherwise interpreted. They may not. It follows directly and unequivocally from the natures of martensite and of austenite, that the change which produces martensite is the allotropic change. If the change which produces martensite is called Ar", and the allotropic change Ar3, then Ar" is identical with Ar3.

The manner in which martensite forms is determined by the natures of martensite and austenite, and there can be no dispute about the validity of the above argument unless the nature of one or other of these constituents is also questioned.

So much emphasis of this point may appear redundant, but there is an evident tendency on the part of some writers to avoid this admission, and to proceed as if the facts would admit of some less definite explanation.

To summarise, martensite is, as far as present experimental knowledge goes, a solid solution of carbon in alpha iron, formed when decomposition of austenite takes place without separation of carbon.

SUB-SECTION B. THE SOLUTE IN MARTENSITE AND AUSTENITE.

In the introduction it was stated that throughout this thesis atomic carbon would be referred to as the solute in all solid solutions occurring in steel. No reason for adopting this nomenclature was given, but in this sub-section an endeavour will be made to show that the implication conveyed by this nomenclature is in accordance with the actual facts.

This sub-section is headed "The Solute in Martensite and Austenite" because it is part of a section devoted to the discussion of these constituents, but what is said with respect to the solute in these constituents applies equally to the solute in

in ferrite, whether the ferrite is free or a constituent of pearlite, troostite or sorbite.

In the section devoted to the consideration of the specific volume results, it was stated that the specific volume of martensitic steels was of the same order as the specific volume of mixtures of alpha iron and graphite. It was further stated that from analogy with specific volume measurements made on various series of non-ferrous solid solutions, this similarity between the specific volume of martensite and mixtures of iron and graphite, might be taken to mean that martensite was a solid solution of graphite (or atomic carbon) in alpha iron. Some current views on various aspects of the specific volume of solid solutions or mixed crystals were cited, and it was shown that there is now a fairly general tendency to look on the solid solutions, which appear in the equilibrium between iron and carbon, as solutions of atomic carbon in one or other of the iron modifications.

In the section devoted to consideration of the dilatation results, it was shown that when mixtures of carbide and ferrite were converted into austenite, the volume changes which accompanied the conversion indicated that the allotropic change took place with a contraction, whereas the solution of carbon took place with an expansion. It was further shown that this expansion caused by the solution of carbon, was proportional to the carbon content, and took effect at the instant of solution. Thus in the eutectoid steel, the volume change observed is the resultant of the contraction due to the allotropic change, and the expansion due to solution of carbon. The whole change takes place at constant temperature, that is, the expansion caused by the carbon takes place simultaneously with the solution of carbon, in the same way as the contraction caused by the allotropic change takes place simultaneously with the allotropic change.

In hyper-eutectoid steels, as the carbon in excess of the eutectoid percentage is dissolved, the expansion which accompanies

accompanies solution causes an increased slope on the dilatation curve, until all the carbon is dissolved.

All this goes to show that solid solutions of carbon in gamma iron have higher specific volumes than mixtures of carbide and gamma iron, and that, if this expansion is brought about by the breakdown of carbide to iron and carbon, the breakdown must take place at the instant of going into solution.

Bragg has said that each atom may, for certain purposes, be regarded as a solid inflexible sphere called the sphere of influence. This view of the atom has been used to a considerable extent by Rosenhain (53) in developing his theories of the properties of alloys, and Thompson (54) in explaining the mechanical properties of steel. So far as non-ferrous solid solutions have been examined, this picture of the function of the atoms has proved satisfactory. It has been made use of in another section of this thesis to support the view that, if solid solutions of carbon in alpha iron have the same specific volume as mixtures of graphite and alpha iron, carbon must be as atomic carbon in those solutions.

But when the conception is carried from solid solutions and applied to intermetallic compounds and allotropic modifications, it no longer explains the observed phenomena.

The specific volume of an intermetallic compound may be notably different from that of a mixture of the constituents in the same proportions, and the specific volume of one allotrope may be notably different from that of another.

In the first case, because the atoms composing the compound have formed a lattice differing from that formed by either of the component atoms alone, the space occupied by the individual atoms has changed. In the second case, because the space lattice of the two allotropes is different, the space occupied by each atom is different.

The space occupied by an atom therefore depends on the

the lattice into which it is built.

When atoms of a metal A are replaced by atoms of another metal B and the lattice structure is unaltered, the space occupied by each atom of the metal A is the same as when the whole lattice was composed of atoms of A, and the space occupied by each atom of the metal B is the same as if the whole lattice were composed of atoms of B. If B has a higher specific volume than A, the specific volume of the solid solution of B in A will increase with increase in the amount of B. But it will not be correct to say - as some writers do - that the lattice of A has been expanded by solution of B, nor to deduce that certain properties are thereby conferred on the metal. All that can be said is that the lattice formed by atoms of A and B is more expanded than a lattice composed only of A, because the atoms of B naturally occupy more space than the atoms of A.

When the atoms of A and B form a lattice differing from that normally formed by either A or B, the space occupied by each atom in the new lattice may be entirely different from that occupied by the same atom in a lattice composed of similar atoms only. This is the state in an intermetallic compound, and it may be suggested that unless this state exists no intermetallic compound exists.

The conception of combination in the solid state, as in intermetallic compounds, must be carefully distinguished from the conception of combination in solution. As it is from consideration of the state of combination in solution that our conception of the molecule has been built up, it is better to avoid all reference to molecules when considering the solid state. There are, in fact, no molecules in the solid state, only lattices built up of atoms of various elements in definite proportions.

If this is true, an inter-metallic compound only exists when its characteristic lattice, formed by a definite amount of each of the constituent atoms, exists.

An inter-metallic compound AB in solution in A, would be characterised by a complex lattice consisting of the lattice peculiar to the compound interpenetrating the lattice peculiar to the pure metal. As far as the writer is aware, such a lattice has not yet been discovered.

An intermetallic compound has no real existence in solution, it exists only in the free state. Solution of an intermetallic compound is accompanied by a breakdown of the compound and a re-arrangement of the atoms composing it.

When a series of solid solutions is interrupted by the appearance of a compound, there is a discontinuous change in properties at the composition corresponding to the compound. This discontinuous change in properties is caused by a sudden change in the space lattice, in the same way as the change in properties which accompanies an allotropic change, is caused by a change in s space lattice.

If an intermetallic compound could exist as such in solution, there is no reason why the change in properties should be discontinuous, for gradually increasing amounts of the compound should produce progressive changes in the properties, in the same way as gradually increasing amounts of one pure metal produce progressive changes in the properties of another, when the two pure metals are soluble in the solid state.

It must therefore be concluded that it is carbon and not carbide which exists in solution in austenite, martensite and ferrite; that the three atoms of iron which were combined with a certain atom of carbon before it went into solution, are not the same atoms which are combined with it when it comes out again; and that the carbon atoms may, when they come out of solution, form cementite or graphite according to the conditions.

SUB-SECTION C. THE TEMPERING OF MARTENSITE.

Some of the specific volume results indicate that the most concentrated solutions of carbon in alpha iron decompose at

at the lowest temperatures. This is in agreement with the conception of martensite as an unstable solution of carbon in a stable modification of iron; the greater the carbon content, the more unstable the solution. The results (Figures 1 and 2) from which these conclusions were drawn, refer only to martensitic chromium steels containing up to .98 per cent of carbon. The steel with 1.73 per cent of carbon was almost entirely austenitic when quenched.

The chromium steels quenched from 1100° C. and immersed in liquid air, constitute a better series of martensitic steels, as the steel with 1.73 per cent of carbon was martensitic after this treatment. When the specific volume of these steels, measured soon after withdrawing from the liquid air, is compared with the specific volume one year afterwards, it will be seen that the steels with the highest carbon content undergo a change even at ordinary temperature.

Specific Volume of Chromium Steels Quenched from 1100° C. and immersed in liquid air.

C	D	E	F	B
.128000	.128780	.129020	.129256	.130150

Specific Volume of above series after one year at ordinary temperature.

C	D	E	F	B
.128000	.128780	.128981	.129185	.130050

Change in Specific Volume during one year at ordinary temperature

C	D	E	F	B
.0000000	.000000	.000039	.000071	.000100

After tempering for 60 hours at 60° C. the specific volume of this series was as follows.

C	D	E	F	B
.128000	.128784	.128943	.129056	.129924

Change in specific volume during 60 hours at 60° C.

C	D	E	F	B
.0000000	.000000	.000038	.000129	.000126

These results show clearly that the statement made at the beginning of this sub-section is correct.

It has been said that these results are in accordance with the view that martensite is a uniform super-saturated solid solution of carbon in alpha iron, which decomposes more readily the greater the degree of super-saturation.

Such a solid solution would decompose, by gradual separation of carbon from the whole mass of the solution, until the carbon concentration reached a critical amount, after which no further change would take place at that temperature.

Microscopic evidence does not support the view that martensite decomposes uniformly at low temperatures: it suggests rather, that the decomposition of the solid solution takes place with the production of troostite in certain areas.

If this is what actually occurs, it is difficult to see why the process of decomposition, once commenced, does not proceed to completion at the rate at which it started. If one area can change, why do others remain unchanged?

In section 4 it is shown that when martensite is tempering at 240° C., the amount of change which takes place in a given time interval is proportional to the carbon content. This applies equally to carbon, chromium, nickel, and nickel-chromium steels.

These results do not contradict the specific volume results. Taken together, the two sets of results show that the more concentrated the alpha solid solution is, the lower the temperature at which it decomposes, but when tempering is performed at a higher temperature than the critical tempering temperature for all steels in a series, the amount of change in a given time is proportional to the carbon content.

In section 4 it is also shown that

(1) Martensite tempers much more rapidly than austenite at all temperatures.

(2) The rate of tempering of martensite increases rapidly as

as the tempering temperature is raised.

(3) The special elements in the amounts present in the steels investigated, do not greatly affect the rate of tempering of martensite.

(4) The greater part of the tempering change takes place soon after the tempering temperature is reached.

In section 2 it is shown that the carbon dissolved in the alpha iron does not separate completely from solution, until after tempering for a time at temperatures above 500°C. It is also shown that after tempering for three hours at 240°C., 350°C., 450°C. and 550°C. respectively, the carbon content of the alpha solid solution is still related to the amount of carbon that was retained in solution by quenching.

From the above it is concluded that there is a limiting state to which tempering at a particular temperature proceeds, and that this limiting state is connected in some way with the amount of carbon which was retained in solution by quenching.

SUB-SECTION D. THE TEMPERING OF AUSTENITE.

In section 1 it is shown that when an austenitic steel is heated for some time at 250°C. and cooled, much of the austenite is converted to martensite. This observation is shown to be in agreement with the results obtained by Maurer, Enlund, and Andrew and Honeyman.

It is also shown by the results obtained on tempering at lower temperatures, that martensite tempers more readily and at lower temperatures than austenite. This observation is in agreement with the results obtained by Enlund (4) and Scott (55).

If martensite tempers more readily and rapidly than austenite, how does martensite accumulate from the tempering of austenite at 250°C. It would appear that in the neighbourhood of 250°C., the relative rates of tempering of these two constituents are reversed, so that austenite tempers more rapidly than

than martensite. But in section 4 it is shown that at 240°C. martensite tempers much more rapidly than austenite, and there is no reason to believe that the rates are reversed at 250°C.

It must therefore be concluded that martensite does not form from austenite during tempering at a particular temperature, but during cooling after tempering. This explains why the volume expansion which accompanies the formation of martensite from austenite, has never been observed in the dilatation curves of austenite or semi-austenitic steels. It also shows that the formation of martensite from austenite during tempering, is the same as the formation of martensite during quenching.

From the foregoing it is evident that when austenite is reheated to a temperature at which it can decompose, it breaks down to troostite, and the reaction will proceed as long as the temperature is maintained. The greater the amount of nickel or chromium in the austenite, the higher the temperature to which it must be heated before it can decompose to troostite. That is why steels with a notable proportion of these elements can be reheated to moderately high temperatures without any change taking place in the austenite.

The change from austenite to troostite will not proceed until the viscosity of the material has been lowered sufficiently to permit of some diffusion of carbon. It is the carbon (in conjunction with nickel, chromium, or other special element) that preserves the iron in the gamma state, and some change must take place in the amount of carbon dissolved before the iron change can proceed.

The change from austenite to martensite is not dependent on viscosity in the same way as the change from austenite to troostite. Austenite will change to martensite whenever the transformation tension of the allotropic change overcomes the inhibitory effect of the dissolved carbon. During progressive cooling the transformation tension is continually increasing, therefore the austenite changes more rapidly to martensite than

than martensite changes to troostite. The transformation tension appears to increase progressively as the temperature is lowered below that at which the allotropic change should normally occur. Therefore austenite which is stable at ordinary temperatures, will change to martensite when immersed in liquid air.

Austenite has therefore a range of stability at low temperature; below the temperature at which it can change to troostite, and above the temperature at which it can change to martensite. In most carbon steels there is no appreciable space between those temperatures, but as the amount of carbon or other element is increased, the range of stability is widened.

Of these changes, that which gives rise to martensite is the allotropic change only, that which gives rise to troostite is the allotropic and carbon change taking place simultaneously. The former is the Ar" of Portevin, the latter is the Ar'.

In carbon steels up to 1.2 per cent of carbon, these two changes are not definitely separated with respect to the temperatures at which they can take place. Thus during rapid cooling the Ar" may take place between 350° C. and 250° C., but if the cooling is suddenly stopped the Ar' will proceed. Martensite is obtained in these steels by quenching because the Ar" is encouraged by cooling, whereas the Ar' is not.

In certain special steels the two changes may take place at very different temperatures, so that there is a considerable range between the temperature below which Ar' cannot take place, and the temperature above which Ar" cannot take place.

SUB-SECTION E. TROOSTITE, SORBITE AND PEARLITE.

During the present research a few points have arisen with respect to the constitution of troostite, sorbite and pearlite.

Although these names are commonly supposed to designate certain characteristic structures in a series of mixtures of ferrite and cementite in different states of division, it is apparent from many of the results

results that there is some further difference between these constituents.

If the state of division were the only variable in the series of mixtures between troostite and pearlite, there should be no change in specific volume, electrical resistance or electrode potential, after all the martensite is decomposed. The state of division may influence the electrical resistance results, but it certainly does not affect the specific volume nor electrode potential results.

The electrode potential measurements made on the carbon and chromium steels tempered at rising temperatures, show that tempering is accompanied by a progressive decrease in the amount of carbon in solution. When plotted against tempering temperature the electrode potentials of the various steels lie on a straight line. As is pointed out in section 2, this does not mean that the amount of carbon in solution is a linear function of the tempering temperature, as the electrode potential and carbon content are not linearly related. Nevertheless it indicates that the amount of dissolved carbon changes continuously from martensite to pearlite.

The specific volume results obtained for tempered chrome steels indicate the same gradual change in dissolved carbon. After all the martensite has been decomposed, the specific volume continues to fall gradually as the tempering temperature is raised. As change in state of division would not affect the specific volume, it must be concluded that at each temperature a definite amount of carbon remains in solution in the alpha iron.

The tempering resistance curves, as described in section 4, become parallel to the time axis when the resistance is still a considerable distance above the annealed value, thus indicating that tempering has proceeded to a limiting value at that particular temperature.

From the foregoing it must be concluded that each partic-

whatever heterogeneity there may be in the solid that is forming, diffusion in the liquid will be sufficiently rapid to maintain a uniform distribution of the solute. During the formation of ferrite from austenite, the carbon rejected by the ferrite will diffuse very slowly in the austenite. Heterogeneity will therefore be produced in the austenite. As the rate of diffusion of carbon in ferrite will also be slow, and as the austenite will probably be heterogeneous in the first place, the conditions which govern the amount of carbon in solution in particular areas of ferrite are obviously complex.

The time of heating, the temperature, the rate of cooling, and the previous treatment will therefore have a pronounced effect on the constitution of ferrite in slowly cooled steels.

It is also apparent from the electrode potential results that purer ferrite may be produced by tempering quenched steels for some time at temperatures near the critical point, than is ordinarily produced by clow cooling through the critical point.

The ferrite which appears as a component of pearlite in annealed steels does not vary in composition with the carbon content.

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T A B L E S 1 TO 100.

COMPOSITION OF STEELS USED.

	C.	MN.	P.	S.	SI.	NI.	CR.
			CHROME STEELS.				
C.	0·35	0·24			0·17		1·56
D.	0·62	0·22			0·18		1·67
E.	0·80	0·28			0·20		1·67
F.	0·98	0·28			0·22		1·68
B.	1·73	0·30			0·21		1·65
			CARBON STEELS.				
1.	0·22	0·08			0·15		
2.	0·48	0·18			0·11		
3.	0·71	0·16			0·10		
4.	0·90	0·20			0·14		
5.	1·19	0·18			0·11		
6.	1·61	0·18			0·13		
7.	1·73	0·20			0·08		
			NICKEL STEELS				
2.	0·26	0·27	·014	·025	0·10	3·64	
4.	0·44	0·33	·015	·038	0·12	3·65	
5.	0·67	0·40	·012	·027	0·11	3·64	
6.	0·78	0·41	·013	·025	0·15	3·64	
7.	1·06	0·42	·013	·026	0·15	3·67	
			NICKEL CHROME STEELS.				
1.	0·51	0·22	0·02	0·05	0·126	3·52	1·72
2.	1·28	0·26	0·02		0·150	3·46	1·80
3.	1·50	0·26			0·190	3·46	1·80

NICKEL-CHROMIUM-MOLYBDENUM STEELS COMPOSITION.

	1	2	3	4	5	6
Carbon	.37	.41	.39	.78	.80	.87
Manganese	.25	.37	.33	.48	.55	.56
Silicon	.08	.12	.13	.18	.20	.22
Sulphur	.03	.03	.04	.04	.04	.04
Phosphorus	.022	.022	.024	.022	.023	.022
Nickel	2.02	3.53	5.02	2.04	3.42	4.52
Chromium	1.42	1.67	1.59	1.61	1.68	1.69
Molybdenum	.78	.89	.87	.87	.86	.87

EFFECT OF COMMON IMPURITIES ON SPECIFIC VOLUME OF STEEL.

The specific volume of iron is raised by the presence of 1 per cent of each of the following elements:

1 per cent of manganese raises specific volume by .00006

1 " " " phosphorous " " " " .00089

1 " " " silicon " " " " .00117

STEEL.		Manganese Correction	Phosphorous Correction	Silicon Correction	Total Correction
Carbon	1.	·0000048	·000009	·000176	··0001898
	2.	·0000108	"	·000129	·0001488
	3.	·0000096	"	·000117	·0001356
	4.	·0000120	"	·000164	·0001850
	5.	·0000108	"	·000129	·0001488
	6.	·0000108	"	·000152	·0001718
	7.	·0000120	"	·000094	·0001150
Nickel	2.	··0000162	·000009	·000117	·0001422
	4.	·0000198	"	·000140	·0001688
	5.	·0000240	"	·000129	·0001620
	6.	·0000246	"	·000176	·0002096
	7.	·0000252	"	·000176	·0002102
Chrome	C.	··0000144	·000018	·000199	·0002314
	D.	·0000132	"	·000211	·0002422
	E.	·0000168	"	·000234	·0002688
	F.	·0000168	"	·000258	·0002928
	B.	·0000180	"	·000246	·0002820
Ni-Cr.	B.	·0000132	·000018	·000147	·0001782
	G1.	·0000156	"	·000175	·0002086
	G2.	·0000156	"	·000223	·0002666

SPECIFIC VOLUME OF QUENCHED STEELS.

TREATMENT	1	2	3	4	5	6	7
CARBON STEELS							
Quenched 800° C.	·12755	·12802	·12836	·12852	·12862	·12868	·12887
Heated to 1100° and quenched at 800°C.					·12941	·12999	·13119
Quenched 900°	·12746	·12797	·12853	·12892	·12903	·12905	·12889
Heated to 1100° and quenched 900°					·12933	·12971	·12987
Quenched 1000°	·12747	·12806	·12864	·12893	·12929	·12927	·12889
Heated to 1100° and quenched at 1000°C.					·12938	·13002	·13026
Quenched 1100°		·12810	·12860	·12895	·12931	·12845	·12822

CHROME STEELS.	C	D	E	F	B
Quenched 820°C.	·127999	·128644	·128754	·128858	·129151
Quenched 900°C.	·128085	·128740	·129035	·129118	·129250
Quenched 1000°C.	·128076	·128915	·129092	·129242	·128880
Quenched 1100°C.	·128002	·128834	·129082	·129254	·127387
Annealed	·127606	·127853	·127894	·128000	·128214

NICKEL STEELS	2	4	6	7
Quenched800°C.	·127887	·127984	·128551	·128122
Quenched 900°C.	·127638	·127802	·128025	·128210
Quenched 1000°C.	·127640	·128035	·128620	·128240
Quenched 1100°C.	·127633	·128336	·128531	·128060
Annealed	·127345	·127425	·127674	·127940

TABLE 6.

TREATMENT.	SPECIFIC VOLUME				
	Spec. C	Spec. D	Spec. E	Spec.F	Spec. B
Quenched 1100° C.	·12805	·12889	·12921	·12939	·12733
TEMPERED.					
3 hours at 100° C.	·12800	·12884	·12890	·12897	·12729
3 hours at 150° C.	·12798	·12879	·12878	·12880	·12723
3 hours at 200° C.	·12795	·12864	·12877	·12877	·12725
3 hours at 250° C.	·12793	·12865	·12880	·12904	·12767
3 hours at 300° C.	·12788	·12850	·12865	·12902	·12895
3 hours at 350° C.	·12781	·12827	·12845	·12864	·12881
9 hours at 350° C.	·12781	·12828	·12844	·12859	·12876
3 hours at 400° C.	·12779	·12821	·12834	·12848	·12865
9 hours at 400° C.	·12776	·12819	·12830	·12842	·12860
12 hours at 400° C.	·12777	·12819	·12827	·12842	·12858
3 hours at 450° C.	·12777	·12818	·12828	·12838	·12858
Annealed	·12764	·12788	·12794	·12802	·12823

ELECTRODE POTENTIAL OF CARBON STEELS.

As Rolled	Quen. 1100°	Tempd. 3 hrs. at 240°	Tempd. 6 hrs. at 240°	Tempd. 3 hrs. at 350°	Tempd. 3 hrs. 450°	Tempd. 3 hrs. at 550°	Tempd. 3 hrs. at 650°	STEEL NUMBER.
·529	·504	·511	·517	·520	·535	·536	·537	1
·522	·494	·502	.5035	·509	·516	·526	·527	2
·517	·486	·497	.497	·505	.511	·520	·5215	3
·516	·482	·496	·495	·503	·5105	·519	·5205	4
·512	·465	.492	·490	.499	·507	·514	·514	5
·509	·457	·487	·4875	·494	·503	·511	·510	6
·5085	·456	·487	·487	.4935	·5025	·510	·510	7

STEEL NUMBER	Quen. at 1000°	Tempd. 3 hrs. 200°C.
1	·502	·508
2	·492	·498
3	·485	·491
4	·483	·4885
5	·476	·483
6	·471	·477
7	·470	·476

Steel Number	Heated to 1100° cooled to 850°C., soaked for 3 hrs. and cooled slowly.	Heated to 1100° and soaked at 850° for 6 hrs.
1	·540	·539
2	·528	·527
3	·517	·518
4	·512	·513
5	·509	·511
6	·508	·5105
7	·510	·5105

Electrode Potential of Carbon Steels.
Supplementary Tables.

Steel Number	Annealed 850° for 1 hour.	Quenched 1000° Tempered 3 hrs at 240° C.	Quenched 1100° Tempered 3 hrs. 680° C.	Quenched 1100° Tempered 3 hrs at 680° C. and quenched.
1	·527	·513	·536	·534 5
2	·520	·504	·522	·523
3	·518	·498	·518	·519
4	·516 5	·497	·516 5	·517
5	·515 5	·494	·510	·511
6	·514	·491	·506 5	·506 5
7	·513 5	·590	·505 5	·505 5

	Heated to 1100° cooled to 780° and quenched	Heated to 1100° cooled to 850° and quenched
1	·495	·512
2	·488	·491
3	·482 5	·482
4	·479 5	·483 5
5	·474 5	·468
6	·472 5	·468
7	·483	·469

ELECTRODE POTENTIAL OF CHROME STEELS.

As Rolled	Quen. 1100°	Tempd. 3 hrs. at 240°	Tempd. 6 hrs. at 240°	Tempd. 3 hrs. at 350°	Tempd. 3 hrs. at 450°	Tempd. 3 hrs. at 550°	Tempd. 3 hrs. at 650°	STEEL NUMBER
·518	·495	·510	·511	·5135	·517	·5225	·524	C
·508	·485	·499	·501	·5025	·511	·516	·518	D
·507	·482	·497	·496	·501	·509	·514	·515	E
·506	·477	·494	·491	·498	·5075	·5115	·517	F
·506	·451	·473	·482	·4905	·500	·509	·5135	B

STEEL NUMBER	Quen. at 1000°	Tempd. 3 hrs. 200°C.
C	·496	·502
D	·483	·490
E	·481	·4875
F	·477	·483
B	·464	·4685

STEEL NUMBER	Soaked for 4½ hrs. at 850°C and slowly cooled	Heated to 1100° cooled to 850° soaked for 3 hrs. and cooled	Heated to 1100° and soaked at 850°C for 4½ hours.
C	·5180	·5195	·522
D	·508	·510	·5105
E	·507	·507	·507
F	·5105	·507	·508
B	·507	·502	·507

Electrode Potential of Chrome Steels.

Supplementary Tables.

Steel Number.	Quenched 1100°C. Tempered 3 hours at 730°C.	Quenched 1100°C. Tempered 3 hours at 730° and Quenched.
C	·527	·525
D	·520	·519
E	·518	·516.5
F	·517	·516.5
B	·513	·508

	Soaked at 850° for 1 hour.	Soaked at 850° for 3 hours.	Soaked at 850° for 6 hours.	Heated to 1100°. Soaked at 850° for 6 hours.
C	•518	·523	·518	·519
D	•508	·513	·509	·509
E	·506	·517	·507	·510
F	·504	·519	·507	·510
B	·502	·515	·508	·511

	Quenched 1000°C. Tempered 3 hours at 240°C.
C	·511
D	·499
E	·497
F	·495
B	·480

Specific Volume of Chromium Steels Immersed in Liquid Air.

Steel Number	Before Immersion	After Immersion	1 Year After	Tempered 60 hrs. at 60 C.
C	.128000	.128000	.128000	.127981
D	.128840	.128780	.128780	.128794
E	.129080	.129020	.128981	.128943
F	.129254	.129256	.129185	.129056
B	.127387	.130150	.130050	.129924

ELECTRODE POTENTIAL.

Nickel-Chrome Steel.

Steel No.	Annealed	Quenched 1150°
G2	•474	.411

Electrical Resistance Before and After Immersion in Liquid Air

Steel	Before Immersion	After Immersion	2 Months After.
Chrome B	41·91	41·53	41·039
Nickel-Chrome G1	37·47	36·56	36·269
" " G2	46·17	43·15	42·805

Temperature Coefficient of Resistance Before and After Immersion in Liquid Air.

Steel	Before Immersion	After Immersion
Chrome B	•0015878	·0013188
Nickel-Chrome G1	·00134145	·0012246
" " G2	.0016642	.0012159

Specific Volume Before and After Immersion in Liquid Air.

Steel Number	Chrome Steels Quenched 1100° Before Immer.	After Immer.	Chrome Steels Quenched 800 Before Immer.	After Immer.
C	•12800	·12800		
D	•12884	·128780		
E	·12908	·129020		
F	·129254	·129256		
B	·127387	·130150		

Carbon Steels Quenched 1100°

	Before Immersion	After Immersion.
1.	.12747	
2.	.12810	
3.	.12860	
4.	.12895	
5.	.12931	
6.	.12845	
7.	.12822	

EFFECT OF COMMON IMPURITIES ON ELECTRICAL RESISTANCE OF STEEL.

1 per cent of silicon raises resistance of iron 13.5 microhms as determined by Burgess, Gumlech and Le Chatelier.

1 per cent of phosphorous raises resistance of iron by 11.0 microhms as determined by D'Amico.

1 per cent of manganese raises resistance of iron by 5.5 microhms as determined by Barrett, Lang, and Matsushita.

STEEL	CORRECTION.	STEEL	CORRECTION.
Carbon 1.	2.8 microhms	Chromium C.	3.62 microhms
2.	2.8 "	D.	3.65 "
3.	2.58 "	E.	4.24 "
4.	2.94 "	F.	4.51 "
5.	3.26 "	B.	4.49 "
6.	2.80 "		
Ni-Cr. B.	3.13 microhms	Nickel 2.	3.03 microhms
G1.	3.67 "	4.	3.76 "
G2	3.52 "	5.	3.79 "
		6.,	4.41 "
		7.	4.53 "

CARBON STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD. M	SPEC. RES. Microhms.
1.A.	·00097375			
	·00097411			
	·00097310			
	·00097377	·00097368	·02281894	13·976
1.B.	·00095975			
	·00095975			
	·00095924			
	·00095855			
	·00095743	·00095886	·12329750	14·042
1.C.	·00096860			
	·00096789			
	·00096773			
	·00096749	·00096793	·02305726	14·027
2.A.	·0010749			
	·0010743			
	·0010723			
	·0010718			
	·0010723			
	·0010717	·0010729	·0232019	15·629
2.B.	·0010626			
	·0010627			
	·0010629			
	·0010626	·0010627	·0232308	15·518

CARBON STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE	MEAN RES.	MEAN RAD.	SPEC. RES.
2.C.	Ohms	Ohms		Microhms
	·0010675			
	·0010666			
	·0010646			
	·0010640			
	·0010639			
	·0010643	·0010651	·0231601	15·153
2.D.	·0010675			
	·0010675			
	·0010671			
	·0010672	·0010673	·1231601	15·538
2.E.	·0010797			
	·0010790			
	·0010808			
	·0010800	·0010799	·0229027	15·546
3.A.	·0011498			
	·0011419			
	·0011481			
	·0011485	·0011488	·02307052	16·659
3.B.	·0011634			
	·0011627			
	·0011632			
	·0011625	·0011630	·02290260	16·740

CARBON STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
3.C.	·0011527			
	·0011516			
	·0011495			
	·0011498			
	·0011522	·0011512	·02312819	16·736
4.A.	·0012945			
	·0012939			
	·0012940			
	·0012971			
	·0012963			
	·0012969	·0012954	·0231427	18·844
4.B.	·0013060			
	·0013054			
	·0013065			
	·0013066	·0013061	·02297952	18·866
4.C.	·0013484			
	·0013480			
	·0013465			
	·0013465	·0013473	·02233497	18·915
5.A.	·0013779			
	·0013776			
	·0013781			
	·0013785	·0013781	·02233447	19·347

CARBON STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
5.B.	·0013431			
	·0013424			
	·0013434			
	·0013428	·0013429	·0227645	19·216
5.C.	·00133800			
	·00133760			
	·00133740			
	·00133770	·00133770	·0228284	19·195
6.A.	·0014942			
	·0014902			
	·0014881			
	·0014877	·0014901	·02256262	21·133
6.B.	·0014967			
	·0014934			
	·0014902			
	·0014906			
	·0014915	·001494	·02226722	20·875
	·0014912			
6.C.	·0012827			
	·0014826			
	·0014843			
	·0014837	·0014833	·02254263	21·028

CARBON STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
6.D.	·0014981			
	·0014980			
	·0015005			
	·0015001	·0014992	·02195177	20·687

NICKEL STEELS ANNEALED 850° FOT 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
2.A.	·0015800			
	·001584			
	·0015809			
	·0015812	·0015806	·02239233	22·247
2.B.	·0015886			
	·0015873			
	·0015894			
	·0015875			
	·0015894	·0015885	·0223122	22·278
2.C.	·0018019			
	·0018007			
	·0018023			
	·0018015	·0018016	·01989141	22·526
4.A.	·0016972			
	·0016979			
	·0016989			
	·0016959			
	·0016956			
	·0016958	·0016969	·02268648	24·198
4.B.	·0016755			
	·0016744			
	·0016740			
	·0016731	·0016742	·02242266	23·597

NICKEL STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
4.C.	·0016724			
	·0016720			
	·0016718			
	·0016726	·0016724	·02295172	24·137
4.D.	·0016619			
	·0016658			
	·0016652			
	·0016651	·0016654	·02303936	24·188
6.A.	·0019916			
	·0019905			
	·0019926	·0019914	·02187699	27·385
5.A.	·0017924			
	·0017916			
	·0017935			
	·0017938	·0017928	·02264837	25·5226
5.B.	·0017543			
	·0017543			
	·0017545			
	·0017545	·0017544	·02280951	25·1535

NICKEL STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
6.B.	·0020287			
	·0020283			
	·0020343			
	·0020331	·0020311	·02172634	27·738
6.C.	·0020058			
	·0020049			
	·0020000			
	·0019989	·0020024	·02204849	27·764
6.D.	·0021323			
	·0021310			
	·0021277			
	·0021280	·0021297	·02069578	27·578
7.A.	·0019808			
	·0019819			
	·0019843			
	·0019832	0019809		
	·0019809	·0019822	·0223600	27·862
7.B.	·0019976			
	·0019971			
	·0020001			
	·0019974			
	·0019971	·0019978	·02205120	27·813

NICKEL STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
7.C.	·0019487			
	·0019504			
	·0019467			
	·0019468			
	·0019485			
	·0019469			
	·0019475	·0019479	·02244210	27·477
7.D.	·0019739			
	·0019728			
	·0019700			
	·0019724	·0019730	·0223810	27·756

CHROME STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
C.A.	·0013744			
	·0013739			
	·0013668			
	·0013634			
	·0013693			
	·0013688			
	·0013734			
	·0013724			
	·0013767			
	·0013750			
	·0013750			
	·0013747			
	·0013748			
	·0013748	·00137239	·0227925	19·662
C.B.	·0013790			
	·0013802			
	·0013795			
	·0013798			
	·0013810			
	·0013804			
	·0013806			
	·0013806			
	·0013792	·00138007	·02288688	19·854

CHROME STEELS ANNEALED 850° FOR ½ HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
C.C.	·0013929			
	·0013906			
	·0013919			
	·0013955			
	·0013938			
	·0013919			
	·0013913	·0013927	·0223829	19·594
C.D.	·0014140			
	·0014132			
	·0014127			
	·0014126	·0014131	·0221803	19·702
D.A.	·0014156			
	·0014156			
	·0014150			
	·0014150	·0014153	·02293505	20·4034
D.B.0	·0014328			
	·0014328			
	·0014326			
	·0014326	·0014327	·02274607	20·4843

CHROME STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
E.A.	·0014887			
	·0014890			
	·0014871			
	·0014871			
	·0014861			
	·0014867	·0014873	·02277992	21·2965
E.B.	·0014765			
	·0014765			
	·0014762			
	·0014762	·0014764	·02293745	21·2865
F.A.	·0015656			
	·0015647			
	·0015619	·0015639	·02292585	22·537
F.B.	·0015466			
	·0015458			
	·0015403			
	·0015395			
	·0015391			
	·0015381	·0015412	·0236270	22·633
F.C.	·0015796			
	·0015780			
	·0015782			
	·0015748			
	·0015741			
	·0015741	·0015760	·0228605_	22·646

CHROME STEELS ANNEALED 850° FOR 1° HOUR.

NO. F.D.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Ohms
	·0015241			
	·0015227			
	·0015256			
	·0015263			
	·0015230			
	·0015220	·0015239	·02350890	22·519
B.A.	·0016567			
	·0016567			
	·0016567			
	·0016567	·0016567	·02271484	23·6540
B.B.	·0017307			
	·0017307			
	·0017293			
	·0017293	·0017300	·02236021	23·3150

NICKEL CHROME STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
A.A.	·0016018			
	·0026004			
	·0026010			
	·0026010	·0026011	·02030019	33·1904
A.B.	·0023445			
	·0023445			
	·0023445			
	·0023474			
	·0023474			
	·0023474			
	·0023368			
	·0023368			
	·0023368			
	·0023334			
	·0023334			
	·0023334			
	·0023386			
	·0023386			
	·0023386	·0023361	·02237570	32·8566
B.A.	·0019006			
	·0019006			
	·0019006			
	·0019103			
	·0019103			
	·0010103			
	·0018970			

NICKEL CHROME STEELS ANNEALED 850° FOR 1 HOUR.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
B.A.	·0018970			
	·0018970			
	·0018965			
	·0019139	·0019036	·02318441	27·7413
B.B.	·0019808			
	·0020014			
	·0019836			
	·0019955	·0019903	·02116738	26·4814
C.A.	·0029186			
	·0029067			
	·0029025			
	·0029217	·0029126	·02249383	32·7100
C.B.	·0022859			
	·0023008			
	·0022790			
	·0033042			
	·0022922	·0022922	·0232668	33·5230

CARBON STEELS QUENCHED FROM 800°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
1.	·0010541			
	·0021538			
	·0010525			
	·0010522			
	·0010525			
	·0010522	·0010529	·02270836	15·029
2.	·0016055			
	·0016055			
	·0016009			
	·0016009			
	·0016009			
	·0016009	·0016024	·02313830	23·306
3.	·0032325			
	·0021298			
	·0021311			
	·0021301	·0021309	·02206238	29·551
	·0025327			
4.	·0025327			
	·0025325			
	·0025320			
	·0025318	·0025322	·02314033	36·832

CARBON STEELS QUENCHED FROM 800°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
5.	·0028682			
	·0028885			
	·0028628			
	·0028605			
	·0028646			
	·0028642	·0028646	·02231124	40·174
6.	·0029955			
	·0029931			
	·0029956			
	·0029947	·0029947	·02251828	42·388

NICKEL STEELS QUENCHED FROM 800°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
2.	·0017969			
	·0017969			
	·0017928			
	·0017936			
	·0017973			
	·0017968	·0017957	·02232635	25·200
4.	·0021533			
	·0021528			
	·0021535			
	·0021532			
	·0021561			
	·0021504	·0021532	·02255286	30·523
5.	·0026653			
	·0026678			
	·0026678			
	·0026617			
	·0026617	·0026649	·02272406	38·0646
6.	·0032439			
	·0032410			
	·0032407			
	·0032406	·0032415	·02190387	44·630

NICKEL STEELS QUENCHED FROM800°

NO.	RESISTANCE	MEAN RES.	MEAN RAD.	SPEC. RES.
	Ohms	Ohms		Microhms.
7.	·0031188			
	·0031203			
	·0031186			
	·0031190	·0031192	·02230044	43·724

CHROME STEELS QUENCHED FROM 800°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
C.	·0023411			
	·0023397			
	·0023409			
	·0023409	·0023406	·02271468	33·419
D.	·0025664			
	·0025664			
	·0025664			
	·0025664			
	·0025667			
	·0025667	·0025665	·02302749	37·1486
E.	·0028381			
	·0028381			
	·0028431			
	·0028431			
	·0028278			
	·0028278			
	·0028218			
	·0028218			
	·0028380			
	·0028380	·0028338	·02286908	40·7353
B.	·0026748			
	·0026733			
	·0026696			
	·0026679			
	·0026675			

CHROME STEELS QUENCHED FROM 800°

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
F.	·0026674	·0026701	·02287954	38·400
B.	·0032002			
	·0032002			
	·0031909			
	·0031909			
	·0032062			
	·0032062			
	·0031968			
	·0031968	·0031985	·02280381	45·8466

NICKEL CHROME STEELS QUENCHED FROM 800°

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
A.	•0040880			
	•0040880			
	•0040727			
	•0040727			
	•0041034			
	•0041034			
	•0040861			
	•0040861	•0040875	•0203874	52•3844
B.	•0028080			
	•0028080			
	•0028086			
	•0028086			
	•0028067			
	•0028067	•0028077	•02320519	40•9535
C.	•0034390			
	•0034390			
	•0034437			
	•0034437			
	•0034355			
	•0034355	•0034394	•02338126	50•5480

CARBON STEELS QUENCHED FROM 900°.

NO.	RESISTANCE	MEAN RES.	MEAN RAD.	SPEC. RES.
	Ohms	Ohms		Microhms
1.	·0010470			
	·0010471			
	·0010470			
	·0010454	·0010466	·02325265	15·297
2.	·0015037			
	·0015023			
	·0015019			
	·0015000			
	·0014954			
	·0014949			
	·0015033			
	·0015026	·0015023	·02323958	21·945
3.	·0021310			
	·0021291			
	·0021298			
	·0021285	·0021296	·02293751	30·704
4.	·0024689			
	·0024682			
	·0024685			
	·0024660			
	·0024652			
	·0024651	·0024665	·02345809	36·396

CARBON STEELS QUENCHED FROM 900°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
5.	·0030386			
	·0030380			
	·0030394			
	·0030382	·0030385	·02292029	43·776
6.	·0034788			
	·0034777			
	·0034816			
	·0034784	·0034791	·02234210	48·859

NICKEL STEELS QUENCHED FROM 900°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
2.	·0017756			
	·0017757			
	·0017726			
	·0017728			
	·0017740			
	·0017744	·0017740	·02229611	24·862
4.	·0021213			
	·0021207			
	·0021194			
	·0021194	·0021202	·02306973	30·745
5.	·0025667			
	·0025667			
	·0025684			
	·0025684			
	·0025684			
	·0025684	·0025678	·02287096	36·9147
6.	·0032236			
	·0032234			
	·0032188			
	·0032095	·0032213	·02180365	44·148

NICKEL STEELS QUENCHED FROM 900°.

NO.	RESISTANCE	MEAN RES,	MEAN RAD.	SPEC. RES.
	Ohms	Ohms		Microhms.
7.	·0035143			
	·0035136			
	·0035101			
	·0035084			
	·0035077			
	·0035077	·0035103	·02220554	48·996

CHROME STEELS QUENCHED FROM 900°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
C.	·0019200			
	·0019190			
	·0019170			
	·0019167	·0019182	·02286612	27·571
D.	·0025858			
	·0025858			
	·0025808			
	·0025808			
	·0025874			
	·0025874	·0025847	·02281770	37·0710
E.	·0028718			
	·0028718			
	·0028687			
	·0028687			
	·0028711			
	·0028711	·0028705	·02302964	41·5527
F.	·0031690			
	·0031697			
	·0031613			
	·0031604			
	·0031692			
	·0031678	·0031662	·02345800	46·686

CHROME STEELS QUENCHED FROM 900°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
B.	·0033895			
	·0033895			
	·0033954			
	·0033954			
	·0033733			
	·0033733			
	·0033821			
	·0033821	·0033851	·02230851	47·4675

NICKEL CHROME STEELS QUENCHED FROM 900°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
A.	·0041963			
	·0041963			
	·0041922			
	·0041922			
	·0041943			
	·0041943	·0041943	·02266835	59·7634
B.	·0028428			
	·0028428			
	·0028403			
	·0028403			
	·0028421			
	·0028421	·0028417	·02281429	40·7521
¢.	·0040697			
	·0040697			
	·0040582			
	·0040582			
	·0040552			
	·0040669			
	·0040669			
	·0040596			
	·0040596	·0040619	·02256539	57·6140

CARBON STEELS QUENCHED FROM 1000°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
1.	·0010598			
	·0010598			
	·0010583			
	·0010573			
	·0010574			
	·0010568	·0010582	·02324537	15·462
2.	·0014988			
	·0014979			
	·0014965			
	·0014990	·0014980	·02319027	21·936
3.	·0020349			
	·0020340			
	·0020363			
	·0020357			
	·0020369			
	·0020364	·0020357	·02317072	29·649
4.	·0025622			
	·0025615			
	·0025601			
	·0025587	·0025606	·02240513	36·062
5.	·0029658			
	·0029651			
	·0029625			
	·0029607	·0029635	·02286875	42·600

CARBON STEELS QUENCHED FROM 1000°.

NO.	RESISTANCE	MEAN RES.	MEAN RAD.	SPEC. RES.
	Ohms	Ohms		Microhms.
6.	·0026929			
	·0036886			
	·0036921			
	·0036909	·0036911	·02246898	52·131

NICKEL STEELS QUENCHED FROM 1000°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
2.	·0020251			
	·0020231			
	·0020221			
	·0020212	·0020229	·02014078	25·610
4.	·0021030			
	·0021017			
	·0021024			
	·0021018	·0021022	·02302114	30·420
5.	·0026124			
	·0026124			
	·0026108			
	·0026108			
	·0026117			
	·0026117	·0026116	·02268256	46·8764
6.	·0030747			
	·0030737			
	·0030739			
	·0030754	·0030744	·02201713	42·548
7.	·0033778			
	·0033744			
	·0033802			
	·0033781	·0033776	·02250672	47·783

CHROME STEELS QUENCHED FROM 1000.°

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
C.	·0019693			
	·0019680			
	·0019671			
	·0019665	·0019677	·02216574	27·416
D.	·0025141			
	·0025141			
	·0025124			
	·0025124			
	·0025168			
	·0025168	·0025144	·02304096	36·4159
E.	·0029132			
	·0029132			
	·0029143			
	·0029143			
	·0029242			
	·0029242	·0029172	·02289710	41·9857
F.	·0032327			
	·0032208			
	·0032255			
	·0032178			
	·0032403			
	·0032409	·0032277	·02287844	46·417

CHROME STEELS QUENCHED FROM 1000°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
B.	·0043034			
	·0043034			
	·0042995			
	·0042995			
	·0043068			
	·0043068	·0043032	·02259188	61·1080

NICKEL CHROME STEELS QUENCHED 1000°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
A.	·0052134			
	·0052134			
	·0052193			
	·0052193			
	·0052127			
	·0052127	·0052152	·02008653	65·8461
B.	·0028088			
	·0028088			
	·0028066			
	·0028066			
	·0028043			
	·0038043	·0028064	·02321353	40·9520
C.	·0039111			
	·0039111			
	·0038915			
	·0038915			
	·0039185			
	·0039185			
	·0039298			
	·0039298			
	·0038988			
	·0038988	·0039099	·02328653	57·2300

CARBON STEELS QUENCHED FROM 1100°.

NO.	RESISTANCE	MEAN RES.	MEAN RAD.	SPEC. RES.
1.	Ohms	Ohms		Microhms.
	·0010724			
	·0010724			
	·0010721			
	·0010726	·0010724	·02272270	15·317
2.	·0015257			
	·0015254			
	·0015257			
	·0015258	·0015256	·02284355	21·905
3.	·0019889			
	·0019880			
	·0019852			
	·0019859	·0019870	·02312437	28·882
4.	·0024035			
	·0024028			
	·0023956			
	·0023962			
	·0024042			
	·0024034	·0024009	·02310795	34·873
5.	·0030094			
	·0030101			
	·0030112			
	·0030118	·0030106	·02238329	42·358

C CARBON STEELS QUENCHED FROM 1100°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
6.	·0038073			
	·0038068			
	·0038069			
	·0038070	·0038070	·02196923	52·572

NICKEL STEELS QUENCHED FROM 1100°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
2.	·0018176			
	·0018169			
	·0018146			
	·0018140	·0018158	·02231778	25·473
4.	·0021175			
	·0021175			
	·0021175			
	·0021173	·0021174	·02306214	30·694
5.	·0024599			
	·0024599			
	·0024616			
	·0024616			
	·0024584			
	·0024584	·0024599	·02287251	35·3660
6.	·0032353			
	·0032339			
	·0032303			
	·0032309	·0032326	·02073928	42·141
7.	·0033303			
	·0033289			
	·0033283			
	·0033289	·0033291	·02245925	46·998

CHROME STEELS QUENCHED FROM 1100°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms
C.	·0019201			
	·0019195			
	·0019162			
	·0019162			
	·0019160			
	·0019152	·0019172	·02258086	27·212
D.	·0024402			
	·0024402			
	·0024408			
	·0024408			
	·0024458			
	·0024458	·0024423	·02271915	34·8776
E.	·0027480			
	·0027480			
	·0027498			
	·0027498			
	·0027494			
	·0027494	·0027491	·02321567	40·1076
F.	·0031417			
	·0031423			
	·0031423			
	·0031433	·0031424	·02351482	46·447

CHROME STEELS QUENCHED FROM 1100°.

NO.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
B	·0037482			
	·0037482			
	·0037569			
	·0037569			
	·0037376			
	·0037376			
	·0037426			
	·0037426	·0037463	·02233979	52·6060

NICKEL CHROME STEELS QUENCHED FROM 1100°.

No.	RESISTANCE Ohms	MEAN RES. Ohms	MEAN RAD.	SPEC. RES. Microhms.
A.	·0045488			
	·0045488			
	·0045469			
	·0045469			
	·0045521			
	·0045521	·0045493	·02252939	64·4240
B.	·0028016			
	·0028000			
	·0028000			
	·0028026			
	·0028026	·0028014	·0227402	41·0023
C.	·0041062			
	·0041062			
	·0041003			
	·0041003			
	·0041114			
	·0041114	·0041059	·0226 020	58·3536

ELECTRICAL RESISTANCE OF CARBON STEELS.

TREATMENT	RESISTANCE (microhms) No 1	No 2	No 3	No 4	No 5	No 6
Quenched 1100° C.	15·34	21·93	28·88	34·87	42·36	52·57
Quenched 1000° C.	15·46	21·94	29·65	36·06	42·60	52·13
Quenched 900° C.	15·30	21·94	30·70	36·39	43·78	48·86
Quenched 800° C.	15·03	23·31	29·55	36·83	40·17	42·39
Tempered at 240° C. after quenching from:-						
1100° C.	15·12	17·73	19·05	22·59	22·69	24·37
1000° C.	15·14	17·47	19·00	22·53	22·75	24·57
900° C.	—	—	19·17	21·48	22·94	25·22
Annealed	14·02	15·56	16·73	18·88	19·13	21·17

ELECTRICAL RESISTANCE OF NICKEL STEELS.

Resistance (microhms)

TREATMENT	Ni3	Ni4	Ni5	Ni6	Ni7
Quenched 1100° C.	25·47	30·69	35·37	42·14	47·00
Quenched 1000° C.	25·61	30·42	36·88	42·55	47·78
Quenched 9000° C.	24·86	30·74	36·91	44·15	48·99
Quenched 800° C.	25·20	30·52	38·06	44·63	43·72
Tempered at 240° C. after quenching from:-	Ni2	Ni4	Ni5	Ni6	Ni7
1100° C.	24·10	26·11	28·02	30·44	31 46
1000° C.	24·54	26·48	28·05	29·98 30	30·91
Annealed	22·35	24·37	25·34	27·62	27·73

ELECTRICAL RESISTANCE OF CHROME STEELS.

Resistance (microhms)

TREATMENT	C	D	E	F	B
Quenched 1100°C.	27·21	34·88	40·10	46·46	52·60
Quenched 1000°C.	27·42	36·42	41·99	46·42	61·11
Quenched 900°C.	27·57	37·07	41·55	46·69	47·47
Quenched 800°C.	33·42	37·15	40·74		45·85
Tempered at 240°C. after quenching from:-					
1100C.	25·44	29·69	33·06	34·32	36·45
1000°C.	25·34	29·81	32·71	33·87	35·40
900°C.	25·64			33·69	
Annealed	19·75	20·45	21·29	22·58	23·99

ELECTRICAL RESISTANCE OF NICKEL-CHROME STEELS.

TREATMENT	Resistance (microhms) B×		G1	G2
Quenched 1100°C.	41·00		58·35	64·42
Quenched 1000°C.	40·95		57·23	65·85
Quenched 900°C.	40·75		57·61	59·76
Quenched 800°C.	40·95		50·55	52·38
Tempered completely at 240°C. after quenching from:-				
1100°C.	35·89		44·44	65·19
1000°C.	35·57		47·56	66·97
Annealed	27·06	32	32·46	32·96

CARBON 1 QUENCHED 1100° C.

Time (mins)	Res. Microhms.	Time	Res.
0	28·40	21	30·55
½	29·50	22	30·53
1	30·53	23	30·52
1½	31·08	24	30·52
2	31·35	25	30·50
2½	31·30	30	30·43
3	31·25	35	30·37
3½	31·12	40	30·28
4	31·04	45	30·25
4½	31·00	50	30·23
5	30·85	55	30·19
5½	30·83	60	30·19
6	30·80	70	30·18
6½	30·79	80	30·18
7	30·79	90	30·17
7½	30·78	100	30·16
8	30·77	110	30·16
8½	30·75	120	30·16
9	30·75	140	30·15
9½	30·73	160	30·13
10	30·73	180	30·12
11	30·71	200	30·11
12	30·70	220	30·11
13	30·70	240	30·09
14	30·67	260	30·08
15	30·65		
16	30·62	Completely	
17	30·62	Tempered...25·41	
18	30·60		
19	30·57		
20	30·56		

CARBON 2 QUENCHED 1100° C.

C2 1100°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	35·95	21	32·43	250	31·79
½	35·48	25	32·39	260	31·77
1	34·90	30	32·34		
1½	34·73	35	32·28	Completely	
2	34·42	40	32·23	Tempered..28·77	
2½	34·21	45	32·20		
3	34·01	50	32·19		
3½	33·79	55	32·17		
4	33·68	60	32·17		
4½	33·57	65	32·15		
5	33·46	70	32·13		
5½	33·40	75	32·12		
6	33·37	80	32·12		
6½	33·32	85	32·08		
7	33·26	90	32·08		
7½	33·20	95	32·06		
8	33·16	100	32·05		
8½	33·10	110	32·02		
9	33·06	120	32·01		
9½	33·00	130	31·97		
10	32·94	140	31·97		
11	32·82	150	31·94		
12	32·77	160	31·93		
13	32·70	170	31·91		
14	32·63	180	31·90		
15	32·57	190	31·88		
16	32·52	200	31·85		
17	32·49	210	31·84		
18	32·46	220	31·82		
19	32·45	230	31·80		
20	32·44	240	31·79		

CARBON 3 QUENCHED 900°C.

C 3 900°.

Time (mins)	Res. Microhms.	Time	Res.
0	46·48	21	33·95
½	39·80	22	33·90
1	38·45	23	33·85
1½	37·58	24	33·82
2	37·23	25	33·79
2½	36·91	27½	33·79
3	36·53	30	33·75
3½	36·28	35	33·70
4	36·04	40	33·65
4½	35·82	45	33·60
5	35·71	50	33·55
5½	35·58	55	33·55
6	35·49	60	33·50
6½	35·41	65	33·50
7	35·28	70	33·50
7½	35·20	75	33·46
8	35·16	85	33·46
8½	35·08	100	33·41
9	35·02	115	33·41
9½	34·92	125	33·41
10	34·84	133	33·38
11	34·77	149	33·38
12	34·62	170	33·36
13	34·50	182	33·36
14	34·44	201	33·34
15	34·34	216	33·33
16	34·22	250	33·32
17	34·18	302	33·31
18	34·12	320	33·30
19	34·06	Completely	
20	34·00	Tempered...31·05	

CARBON 3 QUENCHED 1000° C.

C3 1000°.

Time (mins)	Res. Microhms.	Time	Res.	
0	44·87	21	33·79	
½	41·25	22	33·76	
1	39·00	23	33·75	
1½	37·47	24	33·72	
2	36·81	25	33·69	
2½	36·48	26	33·67	
3	36·23	27	33·65	
3½	36·02	28	33·64	
4	35·88	29	33·63	
4½	35·73	30	33·62	
5	35·60	35	33·58	
5½	35·48	37½	33·58	
6	35·35	42	33·56	
6½	35·22	47½	33·52	
7	35·10	50	33·48	
7½	35·00	52½	33·48	
8	34·85	55	33·44	
8½	34·76	60	33·44	
9	34·70	65	33·40	
9½	34·64	80	33·36	
10	34·60	91	33·33	
10½	34·52	100	33·35	
11	34·48	120	33·28	
12	34·40	160	33·23	
13	34·30	175	33·23	
14	34·21	218	33·20	
15	34·02	250	33·19	
16	33·97	275	33·18	
17	33·93	295	33·17	
18	33·90	315	33·15	
19	33·87	375	33·11	Completely Tem.31·05
20	33·83			

C3

- CARBON 4 QUENCHED 900° C. C4 900°

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	51·77	21	36·62	324	36·53
½	45·27	22	36·58	340	36·53
1	42·64	23	36·54		
1½	41·36	24	36·51	Completely	
2	40·75	25	36·48	Tempered..33.62	
2½	40·31	26	36·45		
3	39·97	27	36·43		
3½	39·62	28	36·41		
4	39·41	29	36·39		
4½	39·23	30	36·37		
5	39·00	32½	36·33		
5½	38·82	35	36·30		
6	38·71	40	36·24		
6½	38·58	45	36·16		
7	38·48	50	36·08		
7½	38·35	60	36·95		
8	38·25	70	36·86		
8½	38·14	75	36·82		
9	38·02	85	36·79		
9½	37·93	100	36·79		
10	37·81	125	36·75		
11	37·64	140	36·70		
12	37·49	148	36·67		
13	37·34	170	36·62		
14	37·23	180	36·60		
15	37·11	198	36·60		
16	37·01	220	36·60		
17	36·92	231	36·60		
18	36·81	251	36·58		
19	36·73	276	36·56		
20	36·67	302	36·56		

CARBON 4 QUENCHED 1000° C. C4 1000°.

Time (mins)	Res. Microhms.	Time	Res.
0	51·29	21	38·06
½	43·35	22	38·03
1	41·62	23	38·00
1½	40·87	24	37·96
2	40·50	25	37·93
2½	40·22	26	37·89
3	39·99	27	37·83
3½	39·75	28	37·77
4	39·57	29	37·74
4½	39·45	30	37·72
5	39·32	35	37·53
5½	39·23	40	37·44
6	39·11	45	37·40
6½	39·02	50	37·35
7	38·95	55	37·31
7½	38·89	60	37·25
8	38·82	65	37·21
8½	38·75	70	37·18
9	38·68	75	37·16
9½	38·62	80	37·16
10	38·56	85	37·14
11	38·51	100	37·11
12 48	38·48	115	37·10
13	38·44	125	37·08
14	38·38	135	37·06
15	38·34	175	37·06
16	38·30	200	37·04
17	38·25	220	37·04
18	38·21	275	37·02
19	38·15	295	37·02
20	38·09	350	37 01 Completely Tem. 33·62

CARBON 4 QUENCHED 1100°C. C4 1100°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	49·60	21	38·54	185	37·27
½	46·67	22	38·51	195	37·23
1	44·83	23	38·48	205	37·20
1½	43·60	24	38·43	215	37·17
2	42·85	25	38·38	225	37·13
2½	42·32	26	38·34	250	37·10
3	41·75	27	38·29	275	37·06
3½	41·46	28	38·24		
4	41·21	29	38·19	Completely	
4½	40·93	30	38·14	Tempered..33·62	
5	40·76	35	37·96		
5½	40·60	37½	37·90		
6	40·45	40	37·81		
6½	40·32	42½	37·79		
7	40·18	45	37·76		
7½	40·02	47½	37·75		
8	39·94	50	37·73		
8½	39·80	53	37·73		
9	39·71	57½	37·68		
9½	39·60	60	37·64		
10	39·51	67	37·60		
11	39·44	75	37·60		
12	39·31	80	37·54		
13	39·19	90	37·54		
14	39·04	100	37·50		
15	38·95	110	37·46		
16	38·87	116	37·46		
17	38·80	135	37·35		
18	38·73	155	37·32		
19	38·64	165	37·31		
20	38·58	175	37·30		

CARBON 5 QUENCHED 900° C. C5 900°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	59·22	21	40·75	321	38·64
½	57·00	22	40·64	350	38·62
1	54·03	23	40·57	362	38·60
1½	51·51	24	40·51	377	38·60
2	49·52	25	40·47		
2½	48·21	27½	40·39	Completely	
3	46·92	30	40·28	Tempered..33·55	
3½	45·90	32½	40·16		
4	45·43	35	40·01		
4½	45·06	37½	39·86		
5	44·75	40	39·70		
5½	44·35	42½	39·61		
6	44·20	45	39·55		
6½	43·92	50	39·35		
7	43·73	60	39·22		
7½	43·54	70	39·16		
8	43·36	75	39·03		
8½	43·14	100	38·92		
9	42·86	110	38·87		
9½	42·75	120	38·82		
10	42·62	126	38·81		
11	42·44	133	38·80		
12	42·19	151	38·80		
13	41·90	176	38·78		
14	41·63	191	38·72		
15	41·47	220	38·70		
16	41·33	230	38·69		
17	41·20	234	38·69		
18	40·97	250	38·69		
19	40·93	275	38·64		
20	40·85	300	38·64		

CARBON 5 QUENCHED 1100°C.

C5 1100°

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	57·24	21	40·43	240	38·90
½	53·05	22	40·36		
1	49·97	23	40·30	Completely	
1½	47·53	24	40·26	Tempered..33·55	
2	46·00	25	40·22		
2½	44·82	26	40·20		
3	44·32	27½	40·16		
3½	43·81	30	40·09		
4	43·53	32½	40·00		
4½	43·25	35	39·90		
5	43·01	40	39·66		
5½	42·77	42½	39·56		
6	42·56	45	39·48		
6½	42·40	50	39·40		
7	42·17	55	39·31		
7½	41·99	60	39·24		
8	41·80	65	39·21		
8½	41·63	70	39·18		
9	41·51	80	39·16		
9½	41·42	95	39·10		
10	41·30	108	38·98		
11	41·18	120	38·98		
12	41·08	130	38·98		
13	41·02	140	38·96		
14	40·91	150	38·94		
15	40·84	160	38·94		
16	40·77	180	38·94		
17	40·70	195	38·94		
18	40·63	210	38·92		
19	40·56	219	38·92		
20	40·49	230	38·90		

CARBON 6 QUENCHED 900° C.

Time (mins)	Res. Microhms.	Time	Res.
0	66·77	21	46·22
½	57·78	22	46·12
1	55·35	23	45·96
1½	54·25	24	45·85
2	53·58	25	45·75
2½	53·00	27½	45·51
3	52·52	30	45·28
3½	52·04	32½	44·95
4	51·70	35	44·84
4½	51·41	40	44·52
5	50·98	45	44·23
5½	50·73	50	43·96
6	50·49	60	43·58
6½	50·21	70	43·21
7	49·87	90	42·55
7½	49·73	102	42·28
8	49·50	115	42·13
8½	49·37	125	42·02
9	49·16	147	41·93
9½	48·97	155	41·85
10	48·77	198	41·66
11	48·51	214	41·64
12	48·23	225	41·60
13	47·96	249	21·57
14	47·71	260	41·55
15	47·45	273	41·55
16	47·22	311	41·55
17	47·00	349	41·53
18	46·78		
19	46·56	Completely	
20	46·35	Tempered...37.11	

CARBON 6 QUENCHED 1000°C.

C6 1000°

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	71·21	21	48·28	240	40·90
½	64·80	22	48·06	250	40·90
1	62·25	23	47·86	265	40·88
1½	60·49	24	47·72	280	40·84
2	59·32	25	47·53	296	40·84
2½	58·33	26	47·43	325	40·78
3	57·65	27	47·31	350	40·78
3½	57·13	28	47·20		
4	56·56	29	47·08	Completely	
4½	56·17	31	36·84	Tempered..37·11	
5	55·78	32	46·65		
5½	55·24	33	46·53		
6	54·81	34	46·42		
6½	54·39	36	46·14		
7	53·81	40	45·71		
7½	53·05	45	45·11		
8	52·51	50	44·59		
8½	52·08	61	43·56		
9	51·78	70	42·63		
9½	51·50	80	42·23		
10	51·37	90	41·78		
11	50·96	100	41·53		
12	50·55	109	41·45		
13	50·06	120	41·33		
14	49·78	137	41·19		
15	49·52	150	41·08		
16	49·34	167	41·01		
17	49·09	181	40·98		
18	48·92	195	40·95		
19	48·69	210	40·93		
20	48·46	235	40·93		

CARBON 6 QUENCHED 1100° C. C6 1100°

Time (mins)	Res. Microhms.	Time	Res.
0	74·35	21	50·16
½	67·51	22	49·83
1	65·26	23	49·57
1½	63·32	24	49·35
2	61·96	25	49·11
2½	60·75	26	48·82
3	59·73	27	48·69
3½	58·84	28	48·53
4	58·13	29	48·38
4½	57·62	30	48·26
5	57·02	35	47·41
5½	56·51	40	46·58
6	56·09	45	45·83
6½	55·68	50	45·25
7	55·39	60	44·28
7½	55·04	70	43·52
8	54·72	80	42·84
8½	54·51	90	42·49
9	54·22	100	42·21
9½	54·01	110	42·03
10	53·79	140	41·79
11	53·41	160	42·74
12	53·05	180	41·68
13	52·59	200	41·52
14	52·36	220	41·47
15	52·03	230	41·43
16	51·70	240	41·40
17	51·30	250	41·37
18	51·03	260	41·35
19	50·67	306	41·24
20	50·43	Completely Tempered...37.11	

NICKEL 2 QUENCHED 1000° C. N 2 1000°

Time (mins)	Res. Microhms.	Time	Res.
0	39·49	21	38·60
½	39·75	22	38·59
1	40·02	23	38·59
1½	39·80	24	38·57
2	39·69	25	38·56
2½	39·49	26	38·55
3	39·26	27	38·55
3½	39·11	28	38·55
4	39·03	29	38·55
4½	38·87	30	38·53
5	38·78	35	38·50
5½	38·76	40	38·47
6	38·75	45	38·44
6½	38·74	50	38·44
7	38·73	55	38·38
7½	38·73	70	38·32
8	38·71	80	38·32
8½	38·71	98	38·27
9	38·70	120	38·22
9½	38·70	131	38·22
10	38·70	139	38·18
11	38·68	170	38·15
12	38·68	185	38·14
13	38·68	200	38·12
14	38·68	215	38·12
15	38·65	226	38·12
16	38·65	249	38·11
17	38·62	273	38·11
18	38·62	294	38·09
19	38·62	312	38·09
20	38·61	Completely Tempered...35·93	

NICKEL 2 QUENCHED 1100° C. N2 1100:

Time (mins)	Res. Microhms.	Time	Res.
0	39·25	21	37·63
½	39·47	22	37·63
1	39·16	23	37·63
1½	38·28	24	37·64
2	38·02	25	37·63
2½	37·83	26	37·63
3	37·80	30	37·62
3½	37·78	35	37·61
4	37·74	40	37·62
4½	37·72	50	37·59
5	37·70	60	37·59
5½	37·70	75	37·58
6	37·70	104	37·58
6½	37·69	140	37·56
7	37·69	191	37·56
7½	37·68	225	37·56
8	37·68	250	37·55
8½	37·68	275	37·55
9	37·70	Completely	
9½	37·67	Tempered...35·93	
10	37·67		
11	37·67		
12	37·67		
13	37·65		
14	37·66		
15	37·65		
16	37·64		
17	37·64		
18	37·64		
19	37·64		
20	37·64		

NICKEL 4 QUENCHED 1000° C.

N4 1000°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	44·59	21	41·49	230	40·72
½	43·94	22	41·44	247	40·71
1	43·48	23	41·41	255	40·70
1½	43·19	24	41·39	270	40·68
2	42·93	25	41·35	295	40·68
2½	42·79	26	41·32	302	40·66
3	42·58	27	41·30	320	40·65
3½	42·49	28	41·28		
4	42·34	29	41·26	Completely	
4½	42·29	30	41·25	Tempered..38·44	
5	42·22	32½	41·23		
5½	42·17	35	41·20		
6	42·12	40	41·18		
6½	42·06	45	41·16		
7	41·99	50	41·13		
7½	41·93	55	41·11		
8	41·88	60	41·09		
8½	41·82	70	41·06		
9	41·80	75	41·03		
9½	41·78	90	40·98		
10	41·75	100	40·92		
11	41·73	115	40·88		
12	41·71	126	40·85		
13	41·66	145	40·83		
14	41·64	150	40·81		
15	41·61	160	40·80		
16	41·59	174	40·79		
17	41·57	186	40·77		
18	41·55	195	40·76		
19	41·53	201	40·74		
20	41·52	220	40·72		

NICKEL 4 QUENCHED 1100° C. N4 1100°.

Time (mins)	Res. Microhms.	Time	Res.
0	45·98	21	40·98
½	44·69	22	40·96
1	43·78	23	40·92
1½	43·02	24	40·90
2	42·52	25	40·86
2½	42·29	27½	40·81
3	42·09	30	40·74
3½	41·97	35	40·70
4	41·92	40	40·62
4½	41·81	45	40·56
5	41·73	50	40·51
5½	41·69	55	40·47
6	41·62	64	40·43
6½	41·56	70	40·40
7	41·51	76	40·37
7½	41·48	82	40·32
8	41·41	94	40·30
8½	41·34	102	40·26
9	41·30	124	40·18
9½	41·28	140	40·15
10	41·24	169	40·12
11	41·20	190	40·08
12	41·23	204	40·06
13	41·17	225	40·04
14	41·14	249	40·02
15	41·09	260	40·00
16	41·08	Completely	
17	41·06	Tempered...38·44	
18	41·04		
19	41·92		
20	41·00		

NICKEL 5 QUENCHED 1000° C. N5 1000°

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	47·15	21	41·86	210	40·61
½	45·18	22	41·81	220	40·58
1	44·51	23	41·75	230	40·56
1½	44·08	24	41·69	240	40·54
2	43·81	25	41·62	250	40·51
2½	43·68	26	41·57	260	40·50
3	43·49	27	41·55		
3½	43·33	28	41·52	Completely	
4	43·22	29	41·50	Tempered..39·18	
4½	43·11	30	41·48		
5	43·05	40	41·44		
5½	42·97	45	41·41		
6	42·94	50	41·37		
6½	42·86	55	41·32		
7	42·82	60	41·27		
7½	42·77	65	41·23		
8	42·71	70	41·18		
8½	42·65	75	41·15		
9	42·59	80	41·11		
9½	42.54	90	41·88		
10	42·50	100	41·03		
11	42·46	110	40·99		
12	42·40	120	40·96		
13	42·34	130	40·91		
14	42·27	140	40·85		
15	42·20	150	40·81		
16	42·15	160	40·77		
17	42·09	170	40·73		
18	42·04	180	40·69		
19	42·00	190	40·66		
20	41·94	200	40·64		

NICKEL 5 QUENCHED 1100°C.

N5 1100°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	49·42	21	44·13	200	43·08
½	47·49	22	44·10	210	43·06
1	46·50	23	44·08	220	43·06
1½	45·85	24	44·06	230	43·04
2	45·53	25	44·04	240	43·03
2½	45·44	26	44·02	250	43·03
3	45·27	27	44·00	260	43·02
3½	45·13	30	43·96		
4	45·04	35	43·90	Completely	
4½	44·98	40	43·82	Tempered..39·18	
5	44·93	45	43·75		
5½	44·87	50	43·70		
6	44·82	55	43·66		
6½	44·76	60	43·61		
7	44·69	65	43·57		
7½	44·63	70	43·54		
8	44·58	75	43·50		
8½	44·55	80	43·45		
9	44·53	85	43·41		
9½	44·50	90	43·38		
10	44·48	95	43·33		
11	44·46	100	43·30		
12	44·42	110	43·28		
13	44·38	120	43·25		
14	44·35	130	43·22		
15	44·32	140	43·19		
16	44·27	150	43·16		
17	44·23	160	43·14		
18	44·21	170	43·13		
19	44·19	180	43·11		
	444·16	190	43·09		

NICKEL 6 QUENCHED 1000°C.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	56·38	21	47·33	284	44·20
½	54·46	22	47·20	310	44·18
1	52·63	23	47·10		
1½	51·29	24	47·01	Completely	
2	50·57	25	46·92	Tempered..41·87	
2½	50·08	26	46·85		
3	49·83	27	46·79		
3½	49·69	28	46·74		
4	49·54	29	46·66		
4½	49·43	30	46·61		
5	49·36	32½	46·45		
5½	49·28	35	46·23		
6	49·20	37½	46·09		
6½	49·08	40	46·01		
8	49·01	45	45·91		
7½	49·94	50	45·80		
8	48·86	55	45·74		
8½	48·78	60	45·62		
9	48·70	65	45·48		
9½	48·62	70	45·40		
10	48·53	75	45·32		
11	48·41	90	45·18		
12	48·29	100	45·01		
13	48·18	122	44·83		
14	48·06	140	44·67		
15	47·94	159	44·52		
16	47·80	182	44·43		
17	47·69	200	44·36		
18	47·58	232	44·27		
19	47·49	248	44·25		
20	47·44	270	44·23		

NICKEL 6 QUENCHED 1100° C.

N6 1100°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	55·82	21	47·82	215	44·86
½	53·01	22	47·36	239	44·81
1	52·13	23	47·32	258	44·76
1½	51·69	24	47·27	283	44 71
2	51·34	25	47·23		
2½	51·04	26	47·17	Completely	
3	50·86	27	47·08	Tempered..41·87	
3½	50·67	28	47·02		
4	50·50	29	46·90		
4½	50·36	30	46·84		
5	50·22	32½	46·76		
5½	50·03	35	46·64		
6	49·94	37½	46·51		
6½	49·76	40	46·43		
7	49·63	42½	46·36		
7½	49·51	45	46·24		
8	49·39	47½	46·11		
8½	49·25	50	45·98		
9	49·13	55	45·87		
9½	49·02	65	45·80		
10	48·91	70	45·73		
11	48·72	77	45·61		
12	48·51	88	45·42		
13	48·32	99	45·31		
14	48·22	106	45·26		
15	48·04	125	45·12		
16	47·92	147	45·01		
17	47·81	160	44·98		
18	47·69	166	44·96		
19	47·56	181	44·93		
20	47·49	198	44·90		

NICKEL 7 QUENCHED 1000° C.

N7 1000°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	62,88	21	49·45	314	44·81
½	62·52	22	49·26		
1	57·21	23	49·14	Completely	
1½	54·36	24	49·02	Tempered..41·80	
2	53·62	25	48·93		
2½	53·27	26	48·76		
3	53·01	27	48·69		
3½	52·79	28	48·56		
4	52·64	29	48·51		
4½	52·45	30	48·44		
5	52·30	35	48·02		
5½	52·22	40	47·68		
6	52·03	45	47·39		
6½	51·92	50	47·02		
7	51·80	55	46·63		
7½	51·71	64	46·42		
8	51·59	70	46·08		
8½	51·51	81	45·71		
9	51·44	92	45·50		
9½	51·33	101	45·43		
10	51·26	114	45·33		
11	51·09	126	45·26		
12	50·91	141	45·19		
13	50·76	159	45·08		
14	50·58	171	45·01		
15	50·43	186	44·98		
16	50·26	200	44·94		
17	50·09	214	44·90		
18	49·92	245	44·90		
19	49·75	269	44·86		
20	49·63	295	44,83		

NICKEL 7 QUENCHED 1100° C.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	61·87	21	51·22	231	46·11
½	61·79	22	51·11	241	45·93
1	59·23	23	51·03	250	45·84
1½	56·76	24	50·94	266	45·74
2	55·21	25	50·82	275	45·69
2½	54·48	26	50·76	302	45·65
3	54·02	27	50·69	324	45·60
3½	53·78	28	50·60		
4	53·61	29	50·52	Completely	
4½	53·50	30	50·45	Tempered..41·80	
5	53·44	35	50·10		
5½	53·31	40	49·70		
6	53·23	45	49·42		
6½	53·12	50	49·11		
7	53.06	55	48·82		
7½	53·01	60	48·70		
8	52·95	65	48·46		
8½	52·88	70	48·27		
9	52·79	75	48·14		
9½	52·73	86	47·88		
10	52·65	93	47·76		
11	52·55	102	47·49		
12	52·40	115	47·35		
13	52·25	125	47·19		
14	52·08	132	47·10		
15	51·93	145	47·01		
16	51·77	155	46·83		
17	51·64	167	46·72		
18	51·52	185	46·54		
19	51·40	200	46·39		
20	51·28	219	46·30		

CHROME C QUENCHED 900° C.

C 900.°

Time (mins)	Res. Microhms	Time	Res.
0	41·28	21	40·81
½	43·50	22	40·76
1	44·52	23	40·70
1½	45·05	24	40·65
2	44·95	25	40·60
2½	44·79	27½40	40·50
3	44·68	30	40·43
3½	44·32	35	40·34
4	44·03	40	40·22
4½	43·77	45	40·14
5	43·52	50	40·08
5½	43·31	55	39·97
6	43·16	60	39·90
6½	43·01	65	29·86
7	42·81	70	39·81
7½	42·63	85	39·76
8	42·51	95	39·70
8½	42·44	110	39·63
9	42·27	126	39·60
9½	42·14	141	39·56
10	42·03	152	39·53
11	41·82	175	39·53
12	41·71	202	39·50
13	41·60	227	39·48
14	41·51	250	39·48
15	41·38	270	39·45
16	41·26	303	39·42
17	41·14		
18	41·02	Completely	
19	40·96	Tempered ...32.77	
20	40·90		

CHROME D QUENCHED 1000° C.

Time (mins)	Res. Microhms.	Time	Res.
0	51·36	21	46·70
½	51·22	22	46·70
1	50·79	23	46·67
1½	49·64	24	46·65
2	49·06	25	46·60
2½	48·72	26	46·54
3	48·48	27	46·50
3½	48·32	28	46·50
4	48·16	29	46·47
4½	48·04	30	46·42
5	47·93	35	46·36
5½	47·82	40	46·20
6	47·75	45	46·03
6½	47·63	50	45·90
7	47·52	555	45·84
7½	47·47	65	45·66
8	47·42	75	45·45
8½	47·35	90	45·31
9	47·31	100	45·25
9½	47·25	115	45·09
1047	47·19	130	44·93
11	47·12	152	44·66
12	47·04	175	44·54
13	47·01	201	44·45
14	46·97	220	44·40
15	46·90	245	44·35
16	46·83	298	44·28
17	46·79	325	44·16
18	46·77	350	44·09
19	46·75	366	44·05
20	46·72	381	44·02

D 1000°.

Time	Res.
409	43·98
425	43·95
460	43·92

Completely Tempered..33·23

CHROME D QUENCHED 1100° C.

D 1100°

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	49·22	21	44·78	250	43·79
½	49·00	22	44·77	269	43·78
1	47·85	23	44·77	280	43·76
1½	47·18	24	44·75	300	43·76
2	46·67	25	44·75	335	43·74
2½	46·21	26	44·72		
3	45·82	27	44·72	Completely	
3½	45·65	28	44·72	Tempered..33·23	
4	45·53	29	44·70		
4½	45·48	30	44·70		
5	45·42	35	44·61		
5½	45·35	40	44·44		
6	45·30	45	44·36		
6½	45·25	50	44·30		
7	45·22	55	44·25		
7½	45·20	60	44·21		
8	45·18	65	44·16		
8½	45·15	70	44·08		
9	45·11	75	44·05		
9½	45·10	91	44·01		
10	45·08	102	43·97		
11	45·06	124	43·94		
12	45·04	139	43.94		
13	45·03	150	43/93		
14	45·00	163	43·90		
15	44·97	173	43·89		
16	44·93	180	43·89		
17	44·90	195	43·89		
18	44·86	201	43·84		
19	44·83	215	43·82		
20	44·78	231	43·80		

CHROME E QUENCHED 1000° C.

Time (mins)	Res. Microhms	Time	Res.
0	60·06	21	48·78
½	55·25	22	48·73
1	53·48	23	48·69
1½	52·07	25	48·63
2	51·32	27½	48·58
2½	50·86	30	48·52
3	50·56	35	48·45
3½	50·32	40	48·36
4	50.11	45	48.24
4½	49·93	50	48·08
5	49·78	55	47·96
5½	49·62	75	47·73
6	49·57	90	47·61
6½	49·53	110	47·52
7	49·50	125	47·40
7½	49·46	135	47·40
8	49·40	145	47.35
8½	49·37	165	47·30
9	49·32	186	47·26
9½	49·28	200	47·17
10	29·24	221	47·03
11	49·19	251	47·00
12	49·12	279	46·95
13	49·05	Cpmpletely	
14	49·00	Tempered....34·12	
15	48·95		
16	48·92		
17	48·90		
18	48·90		
19	48·85		
20	48·81		

CHROME E QUENCHED 1100° C.

Time (mins)	Res. Microhms.	Time	Res.
0	57·32	21	48·27
½	55·15	22	48·23
1	53·21	23	48·19
1½	52·09	24	48·17
2	51·38	25	48·13
2½	50·78	26	48·09
3	50·33	27	48·03
3½	50·01	28	47·99
4	49·70	29	47·96
4½	49·59	30	47·92
5	49·50	35	47·80
5½	49·42	40	47·72
6	49·33	45	47·69
6½	49·27	50	47·54
7	49·19	55	47·48
7½	49·13	60	47·43
8	49·08	70	47·34
8½	48·95	80	47·25
9	48·91	90	47·16
9½	48·86	100	47·08
10	48·80	110	47·01
11	48·74	120	46·96
12	48·67	130	46·85
13	48·60	140	46·79
14	48·54	150	46·75
15	48·48	160	46·73
16	48·44	170	46·69
17	44·40	180	46·66
18	48·37	190	46·62
19	48·33	200	46·59
20	48·30	210	46·57

E 1100°

Time	Res.
220	46·55
230	46·54
240	46·52
250	46·50
260	46·49
Completely Tempered..	34·12

CHROME F QUENCHED 900° C. F 900.°

Time (mins)	Res. Microhms	Time	Res.
0	62·56	21	50·32
½	62·49	22	50·25
1	61·90	23	50·20
1½	54·75	24	50·14
2	53·70	25	50·06
2½	53·29	26	50·01
3	53·01	27	49·94
3½	52·82	28	49·86
4	52·66	30	49·72
4½	52·52	32½	49·57
5	52·38	35	49·44
5½	52·24	37½	49·25
6	52·12	40	49·09
6½	52·03	42½	49·00
7	51·90	45	48·92
7½	51·78	50	48·74
8	51·69	58	48·58
8½	51·60	65	48·41
9	51·52	70	48·33
9½	51·44	81	48·26
10	51·33	100	48·02
11	51·22	116	47·93
12	51·03	145	47·84
13	50·91	156	47·80
14	50·77	180	47·71
15	50·68	200	47·62
16	50·62	218	47·56
17	50·56	254	47·53
18	50·51	303	47·50
19	50·45	316	47·50
20	50·38	Completely $ Tempered..35·41	

CHROME F QUENCHED 1000°C.

Time (mins)	Res. Microhms.	Time	Res.
0	62·21	21	50·49
½	62·00	23	50·41
1	61·47	24	50·35
1½	55·42	25	50·27
2	54·01	27½	50·09
2½	53·66	30	50·01
3	53·42	35	49·87
3½	53·17	40 49	49·78
4	52·94	45	49·64
4½	52·79	50	49·51
5	52·68	55	49·39
5½	52·52	60	49·30
6	52·41	68	49·21
6½	52·31	78	49·02
7	52·23	90	48·82
7½	52·12	99	48·61
8	52·01	110	48·51
8½	51·90	125	48·47
9	51·78	145	48·41
9½	51·67	171	48·33
10	51·56	200	48·17
11	51·44	230	48·09
12	51·30	261	48·07
13	51·18	286	48·03
14	51·06	316	48·00
15	50·96	326	48·00
16	50·83	342	47·98
17	50·77	359	47·98
18	50·70	372	47·95
19	50·65	Completely	
20	50·57	Tempered...35·41	

CHROME F QUENCHED 1100°C.

F 1100°.

Time (mins)	Res. Microhms.	Time	Res.
0	62·23	21	50·41
½	61·95	22	50·34
1	61·58	23	50·28
1½	56·32	24	50·21
2	54·81	25	50·14
2½	54·25	26	50·09
3	53·83	27	50·03
3½	53·66	28	49·96
4	53·38	29	49·89
4½	53·14	30	49·83
5	52·79	34	49·72
5½	52·55	40	49·46
6	52·37	52	49·13
6½	52·16	66	48·91
7	52·01	75	48·80
7½	51·84	86	48·65
8	51·76	100	48·52
8½	51·59	115	48·43
9	51·48	125	48·38
9½	51·39	140	48·26
10	51·30	151	48·20
11	51·21	169	48·14
12	51·07	180	48·06
13	50·93	210	47·97
14	50·81	225	47·95
15	50·74	236	47·90
16	50·70	263	47·87
17	50·64	288	47·83
18	50·55	330	47·79
19	50·50	351	47·75
20	50·44	381	47·74

Time	Res.
410	47·70

Completely Tempered...35·41

CHROME B QUENCHED 1000 C°.

B1000°

(Time (mins	Res. Microhms	Time	Res.	Time	Res.	Time	Res.
0	82·90	21	76·08	134	68·23	531	52·61
½	80·70	22	76·01	140	67·76	550	52·59
1	79·98	23	75·94	146	67·39	571	52·36
1½	79·48	24	75·85	151	67·04		
2	79·06	25	75·78	163	66·45	Completely	
2½	78·81	26	75·69	173	66·01	Tempered:	
3	78·64	27	75·56	190	65·01	39·61	
3½	78·50	28	75·49	206	64·03		
4	78·35	29	75·42	220	63·18		
4½	78·20	30	75·34	225	62·92		
5	78·03	31	75·26	233	62·61		
5½	77·90	33	75·16	243	62·05		
6	77·77	36	74·85	249	61·55		
6½	77·69	40	74·53	274	60·32		
7	77·60	45	74·16	290	59·53		
7½	77·52	47½	74·02	302	58·96		
8	77·47	50	73·91	308	58·59		
8½	77·38	55	73·50	320	58·04		
9	77·27	60	73·22	335	57·49		
9½	77·16	65	72·85	357	56·62		
10	77·07	70	72·47	363	56·41		
11	76·97	75	72·18	376	55·89		
12	76·78	81	71·72	385	55·57		
13	76·63	90	71·04	392	55·43		
14	76·59	95	70·88	400	55·18		
15	76·53	100	70·47	416	54·65		
16	76·47	105	70·16	424	54·43		
17	76·41	110	69·75	435	54·14		
18	76·34	115	69·49	444	53·93		
19	76·25	120	69·03	476	53·35		
20	76·16	124	68·95	508	52·89		

CHROME B QUENCHED 1100° C.

B 1100°.

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	78·19	21	67·47	234	59·32
½	75·50	222	67·39	245	58·93
1	73·93	23	67·31	260	58·49
1½	72·39	24	67·28	270	58·33
2	71·28	25	67·23	285	57·88
2½	70·52	26	67·17	301	57·43
3	70·03	27	67·09	325	56·83
3½	69·68	28	67·03	350	56·54
4	69·42	29	66·95	375	56·08
4½	69·19	30	66·90	390	55·93
5	69·06	32½	66·85	400	55·82
5½	68·94	35	66·77	432	55·46
6	68·83	40	66·62	449	55·19
6½	68·72	45	66·51	478	54·96
7	68·59	50	66·39	500	54·72
7½	68·51	60	66·12	524	54·50
8	68·44	69	65·87	538	54·40
8½	68·33	75	65·52	560	54·13
9	68·28	85	65·12		
9½	68·23	99	64·48	Completely	
10	68·20	110	64·01	Tempered..39·61	
11	68·12	120	63·72		
12	68·03	130	63·11		
13	67·96	140	62·70		
14	67·90	157	62·14		
15	67·85	165	61·75		
16	67·78	174	61·47		
17	67·72	188	60·81		
18	67·63	200	60·48		
19	67·56	216	59·81		
20	67·52	225	59·52		

CHROMIUM B QUENCHED 1000° C. β 1000°.

(TIME (mins)	RES. Microhms.	TIME	RES.	TIME	RES.
0	83·35	21	76·33	327	58·71
½	81·05	22	76·25	328	58·51
1	80·22	23	76·16	329	58·31
1½	79·74	24	76·05	330	58·22
2	79·36	25	76·00	331	58·14
2½	79·04	30	75·53	332	58·07
3	78·78	35	75·11	333	58·01
3½	78·56	40	74·79	334	57·95
4	78·39	45	74·51	335	57·88
4½	78·25	50	74·24	340	57·57
5	78·06	55	73·85	345	57·33
5½	77·94	60	73·40	350	57·18
6	77·81	70	72·65	355	57·00
6½	77·75	80	71·94	360	56·81
7	77·63	90	71·18	365	56·69
7½	77·52	100	70·53	370	56·55
8	77·46	110	69·77	375	56·47
8½	77·38	130	68·43	390	55·94
9	77·34	150	67·29	400	55·66
9½	77·30	170	66·02	425	55·22
10	77·25	200.	64·12	450	54·81
11	77·20	225	62·63	475	54·54
12	77·13	250	61·14	500	54·49
13	77·05	275	60·13	525	54·37
14	76·93	300	59·22	550	54·22
15	76·81	320	58·68	575	54·11
16	76·75	Withdrawn and replaced		600	54·00
17	76·68	325	59·83	625	53·91
18	76·59	325½	59·25		
19	76·50	326	59·08		
20	76·43	326½	58·83		

NICKEL-CHROME B QUENCHED 1000° C. NCrB 1000°

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	54·82	21	50·38	301	48·93
½	52·79	22	50·35	321	48·93
1	52·43	23	50·33	353	48·92
1½	52·03	24	50·30	404	38·90
2	51·89	25	50·28		
2½	51·54	26	50·26	Completely	
3	51·43	27	50·24	Tempered..27·06	
3½	51·32	28	50·22		
4	51·26	29	50·19		
4½	51·20	30	50·16		
5	51·14	31	50·14		
5½	51·08	35	50·07		
6	51·04	40	50·01		
6½	51·01	45	49·96		
7	50·96	50	49·91		
7½	50·90	61	49·85		
8	50·85	70	49·78		
8½	50·79	82	49·69		
9	50·75	90	49·60		
9½	50·72	101	49·54		
10	50·69	112	49·43		
11	50·66	124	49·37		
12	50·62	140	49·30		
13	50·59	157	49·26		
14	50·56	185	49·21		
15	50·53	199	49·16		
16	50·51	208	49·13		
17	50·48	225	49·07		
18	50·46	240	49·02		
19	50·43	263	48·96		
20	50·41	280	48·94		

NICKEL-CHROME B QUENCHED 1100°C. NCrB 1100.°

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	54·89	21	50·91	378	49·75
½	54·14	22	50·87		
1	53·53	23	50·84	Completely	
1½	53·08	24	50·81	Tempered..27·06	
2	52·82	25	50·78		
2½	52·52	26	50·74		
3	52·35	27	50·72		
3½	52·18	28	50·69		
4	52·01	29	50·67		
4½	51·82	30	50·65		
5	51·76	35	50·53		
5½	51·67	40	50·45		
6	51·58	45	50·38		
6½	51·54	50	50·30		
7	51·50	55	50·25		
7½	51·47	60	50·20		
8	51·40	65	50·20		
8½	51·35	70	50·18		
9	51·33	80	50·14		
9½	51·30	85	50·06		
10	51·25	90	50·04		
11	51·22	100	50·00		
12	51·19	125	49·95		
13	51·15	150	49·90		
14	51·13	175	49·88		
15	51·11	201	49·86		
16	51·09	222	49·83		
17	51·04	249	49·80		
18	51·01	270	49·80		
19	50·97	301	49·78		
20	50·95	352	49·77		

NICKEL-CHROME G1 QUENCHED 1000°C. NCG1 1000°

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	74·91	21	67·95	396	61·23
½	71·53	22	67·92	442	60·89
1	69·94	23	67·90	507	60·48
1½	69·48	24	67·87	555	60·19
2	69·22	25	67·85	601	59·87
2½	69·13	26	67·82	626	59·69
3	69·01	27	67·79		
3½	68·92	28	67·74		
4	68·81	29	67·70	Completely	
4½	68·75	30	67·65	Tempered..32·46.	
5	68·70	35	67·45		
5½	68·67	40	67·25		
6	68.63	45	67·19		
6½	68·60	50	66·93		
7	68·55	55	66·75		
7½	68·51	65	66·29		
8	68·48	73	66·04		
8½	68·46	90	65·72		
9	68·43	102	65·49		
9½	68·39	122	65·15		
10	68·36	140	64·89		
11	68·33	171	64·21		
12	68·29	195	63·75		
13	68·26	218	63·45		
14	68·23	245	63·02		
15	68·20	260	62·83		
16	68·17	275	62·54		
17	68·14	304	62·26		
18	68·10	328	61·91		
19	68·04	350	61·69		
20	68·00	375	61·51		

NICKEL-CHROME G1 QUENCHED 1100°C. NC G1 1100°

Time (mins(	Res. Microhms.	Time	Res.	Time	Res.
0	76·37	21	68·58	226	63·80
½	73·19	22	68·53	253	63·53
1	71·81	23	68·50	260	63·41
1½	71·04	24	68·45	281	62·97
2	70·47	25	68·41	320	62·54
2½	70·10	26	68·37	350	62·35
3	69·81	27	68·34	378	62·01
3½	69·62	28	68·32	400	61·73
4	69·51	29	68·28	427	61·45
4½	69·45	30	68·25	441	61·34
5	69·41	35	68·12	465	61·13
5½	69·33	40	68·00	488	60·92
6	69·27	45	67·94		
7½	69·23	50	67·85	Completely	
8	69·18	55	67·67	Tempered..32·46	
8½	69·15	60	67·52		
9	69·11	65	67·38		
9½	69·07	70	67·19		
10	69·03	75	67·03		
11	69·00	80	66·96		
12	68·97	85	66·85		
13	68·95	95	66·59		
14	68·90	100	66·48		
15	68·85	110	66·23		
16	68·80	120	65·98		
17	68·76	130	65·75		
18	68·71	150	65·45		
19	68·67	170	64·88		
20	68·62	200	64·25		

NICKEL-CHROMIUM G1 QUENCHED 1100° C. NC.G1 1100°

Time (mins)	Res. Microhms.	Time	Res.	Time	Res.
0	76·53	70	67·62	300	62.01
½	74·45	80	67·39	325	61.83
1	72·79	90	67·22	350	61·58
1½	72·36	100	67·03	375	61·31
2	70·74	126	66·51	400	61·13
2½	70·37	150	65·93	425	60·91
3	70·16	175	65·45	450	60·61
4	69·93	200	64·84	Withdrawn and replaced.	
5	69·78	225	64·42	455	61.54
6	69·69	Withdrawn and replaced		456	61.24
7	69·54	230	65·20	457	61.09
8	69·49	231	65·64	458	60.76
9	69·37	232	65·27	459	60.65
10	69·28	233	64·99	460	60.52
11	69·18	234	64·76	465	60.23
12	69·11	235	64·51	470	60.01
13	69·06	236	64·29	475	59.89
14	69·03	267	64·15	480	59.69
15	68·95	238	64·12	485	59.55
17	68·84	239	63·89	490	59.45
20	68·72	240	63·74	495	59.36
25	68·63	245	63·23	500	59.28
30	68·52	250	62·91	505	59.13
35	68·39	255	62·78	555	58.72
40	68·28	260	62·67	605	58.31
45	68·20	265	62·54	655	57.91
50	68·11	270	62·43		
55	67·80	275	62·31		
60	67·71	280	62·24		

NICKEL-CHROME G2 QUENCHED 1000° C.

Time (mins)	Res. Microhms.	Time	Res.
0	90·52	21	92·08
½	91·46	22	92·08
1	91·83	23	92·10
1½	92·01	24	92·10
2	92·05	25	92·07
2½	92·08	27½	92·05
3	92·12	30	92·05
3½	92·12	35	92·07
4	92·12	40	92·05
4½	92·12	45	92·05
5	92·12	50	92·07
5½	92·12	60	92·04
6	92·12	65	92·02
6½	92·12	70	92·03
7	92·12	75	92·03
7½	92·12	100	92·02
8	92·12	120	92·02
8½	92·12	125	92·02
9	92·12	146	92·02
9½	92·12	171	91·98
10	92·12	200	91·99
11	92·12	222	91·99
12	92·12	250	91·99
13	92·12	266	91·98
14	92·12	287	91·97
15	92·12	301	91·97
16	92·08	359	91·96
17	92·09		
18	92·08	Completely	
19	92·08	Tempered...32.96	
20	92·10		

NICKEL-CHROME G2 QUENCHED 1100° C.

Time (mins)	Res. Microhms.	Time	Res.
0	88·53	35	90·06
½	89·56	40	90·04
1	90·11	45	90·06
1½	90·19	55	90·06
2	90·19	60	90·04
2½	90·18	70	90·04
3	90·16	90	90·02
3½	90·17	100	90·00
4	90·16	110	90·00
4½	90·17	125	89·97
5	90·16	150	89·99
5½	90·17	160	89·93
6	90·15	185	89·93
6½	90·15	200	89·90
7	90·15	250	89·85
7½	90·13	275	89.85
8	90·15	300	89·82
8½	90·15	325	89·79
9	90·14	350	89·77
9½	90·14	400	89·73
10	90·13	425	89·70
11	90·13	450	89·67
14	90·12	475	89·64
16	90·12		
18	90·10	Completely	
20	90·10	Tempered...32·96	
25	90·06		
30	90·06		

TEMPERATURE COEFFICIENT OF RESISTANCE.

CARBON STEELS.

NUMBER	TEMPERATURE COEFFICIENT	
	ANNEALED	QUENCHED
1	•004770	•0038077
2	•004220	•0028059
3	•0038012	•0022818
4	•0034759	•0018790
5	•0033511	•0015654
6	•0033575	•00162765

NICKEL-CHROME STEELS.

NUMBER	QUENCHED	TEMPERED
B	•0015028	•0017225
G1	•00134145	•0014962
G2	•0016642	•0016642

TEMPERATURE COEFFICIENT OF RESISTANCE.

CHROME STEELS.

NUMBER	TEMPERATURE COEFFICIENT	
	ANNEALED	QUENCHED
C	·0029310	·0022129
D	·0026930	·0018249
E	·0025924	·0019151
F	·0025316	·0013295
B	.0028420	·0015878

NICKEL STEELS

NUMBER	ANNEALED	QUENCHED
2	·0026970	·0024021
4	·0025761	·0020667
5	·0024302	·0017326
6	·0022909	·0014437
7	·0022560	·0014048

FIG. 1

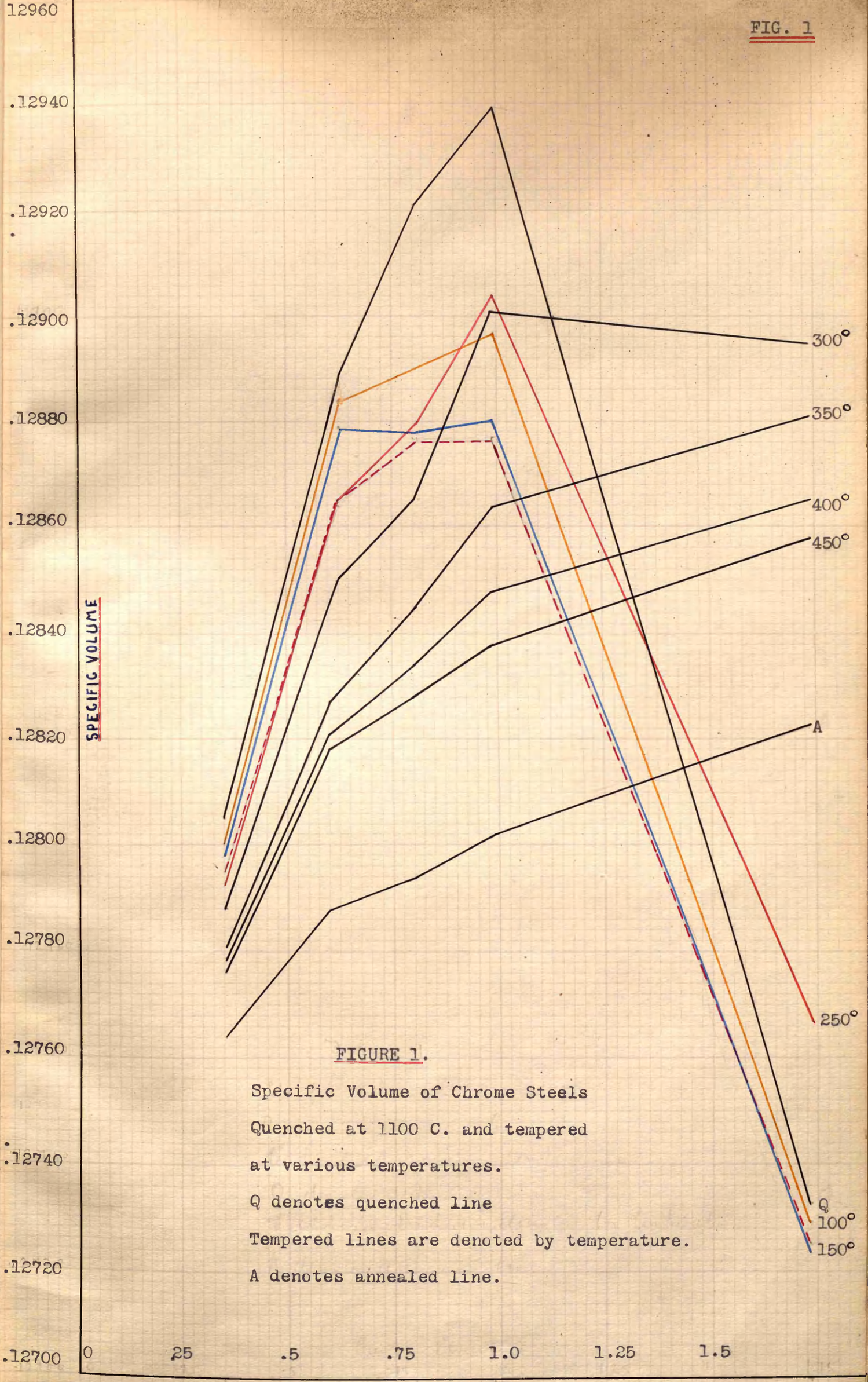


FIGURE 1.

Specific Volume of Chrome Steels
Quenched at 1100 C. and tempered
at various temperatures.
Q denotes quenched line
Tempered lines are denoted by temperature.
A denotes annealed line.

FIG. 2.

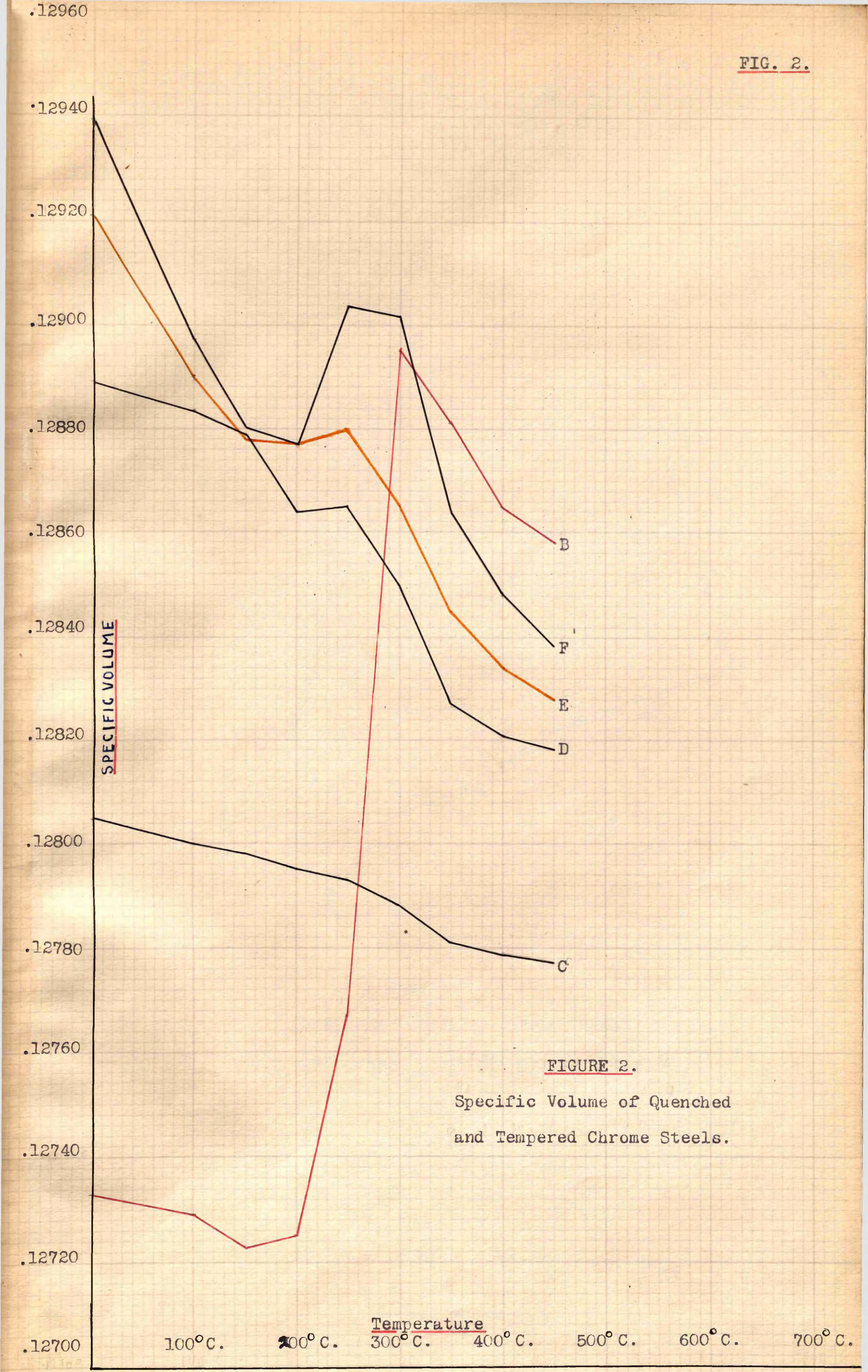


FIGURE 2.

Specific Volume of Quenched and Tempered Chrome Steels.

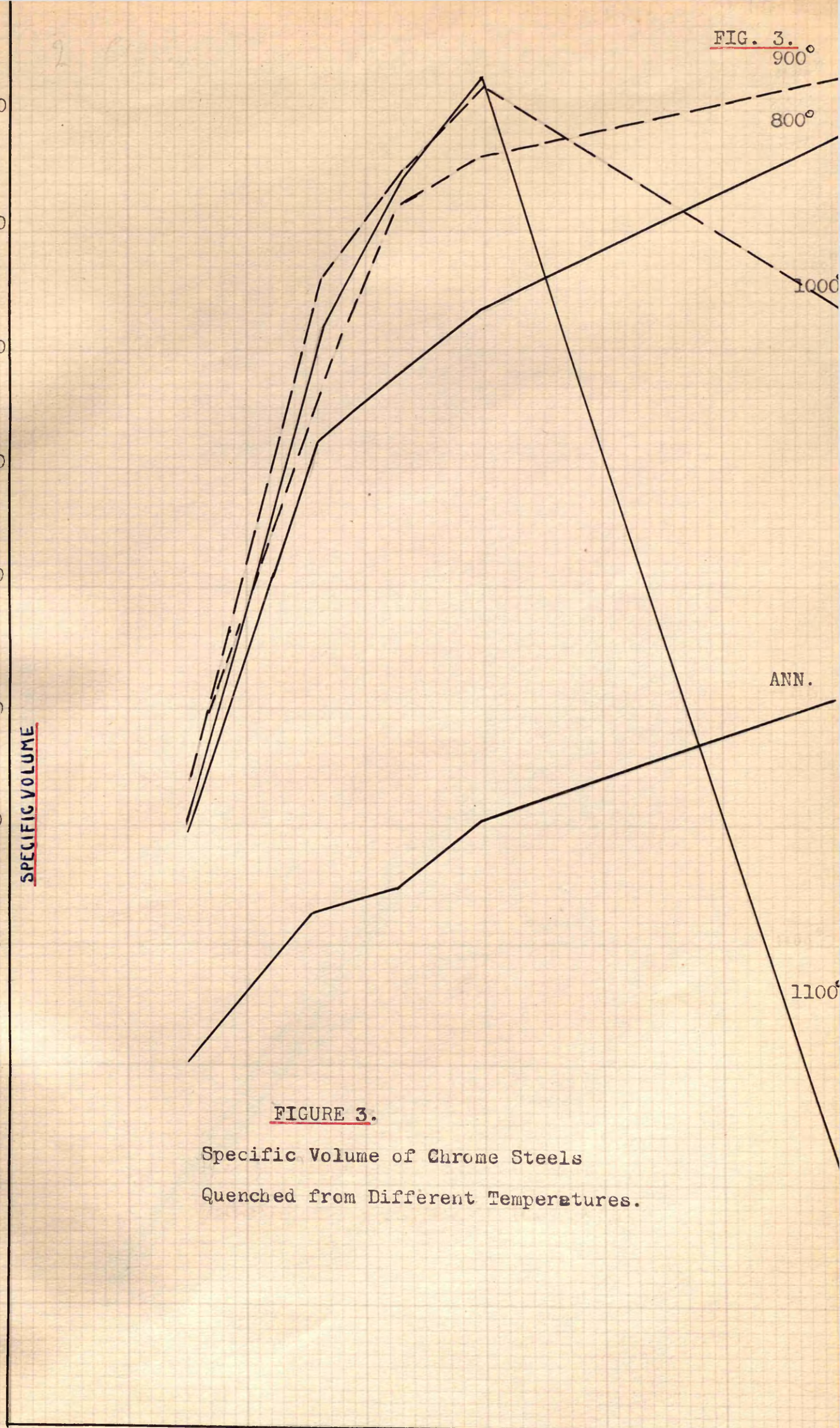


FIGURE 3.

Specific Volume of Chrome Steels

Quenched from Different Temperatures.

FIGURE 4.

Specific Volume of Carbon Steels Quenched from Different Temperatures.

Curve A Heated to 1100° C. Cooled in furnace to 800° C. and quenched.

Curve B Heated to 1100° C. Cooled in furnace to 900° C. and quenched.

Curve C Heated to 1100° C. Cooled in furnace to 1000° C. and quenched.

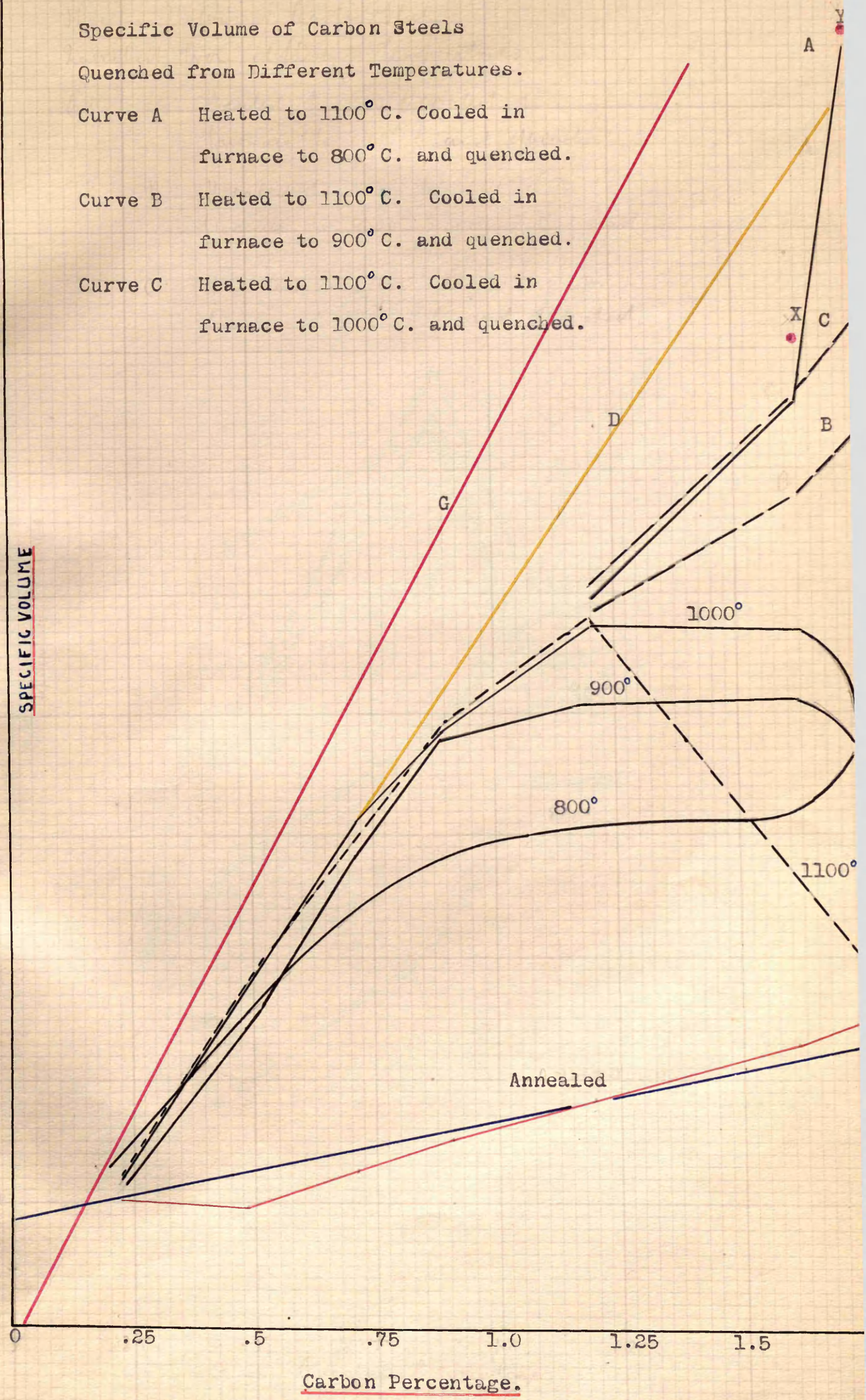


FIG. 5.

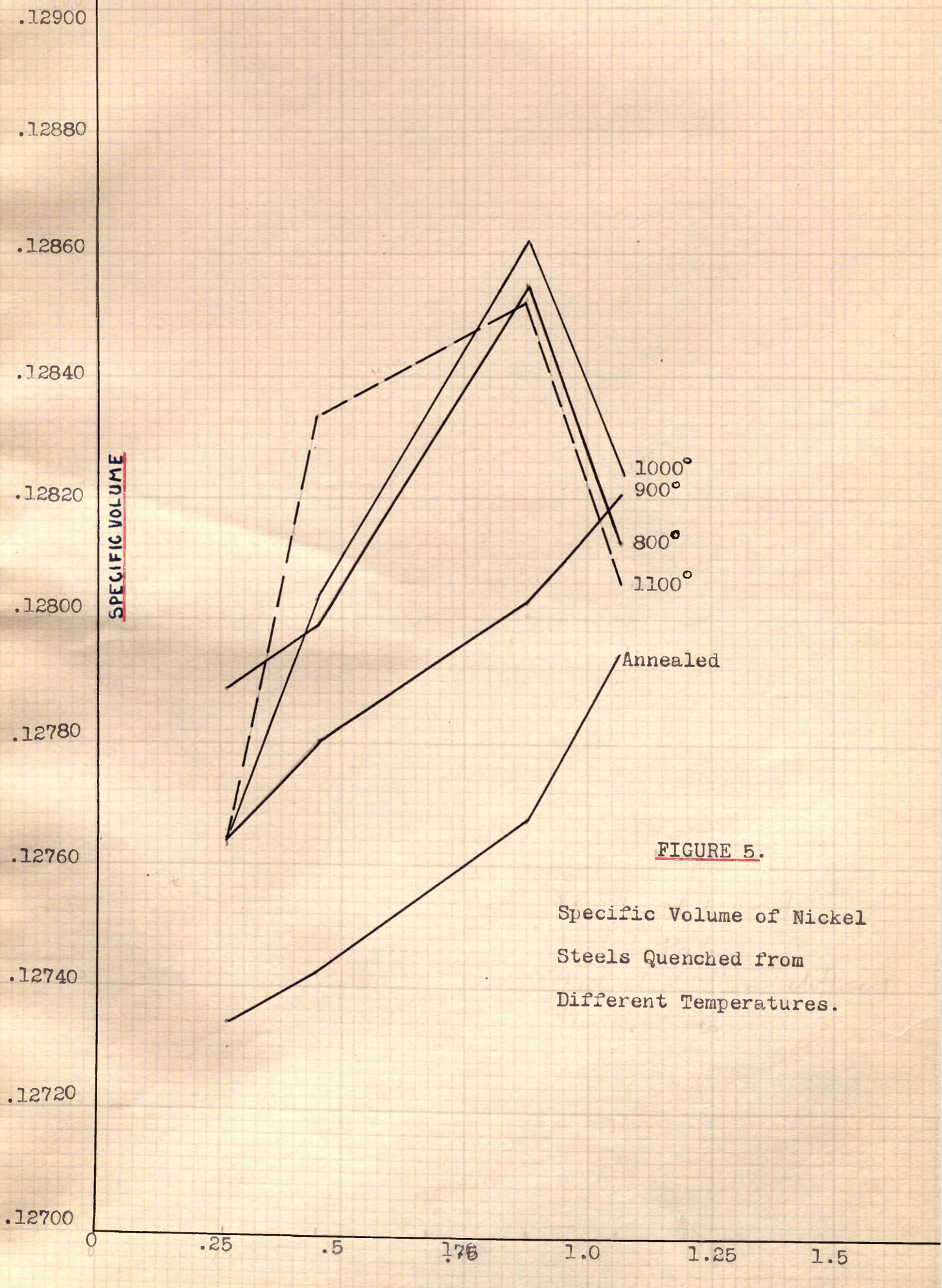


FIGURE 5.

Specific Volume of Nickel Steels Quenched from Different Temperatures.

FIGURE 6.

Showing the effect of Austenite
on the Specific Volume of Chromium Steels.

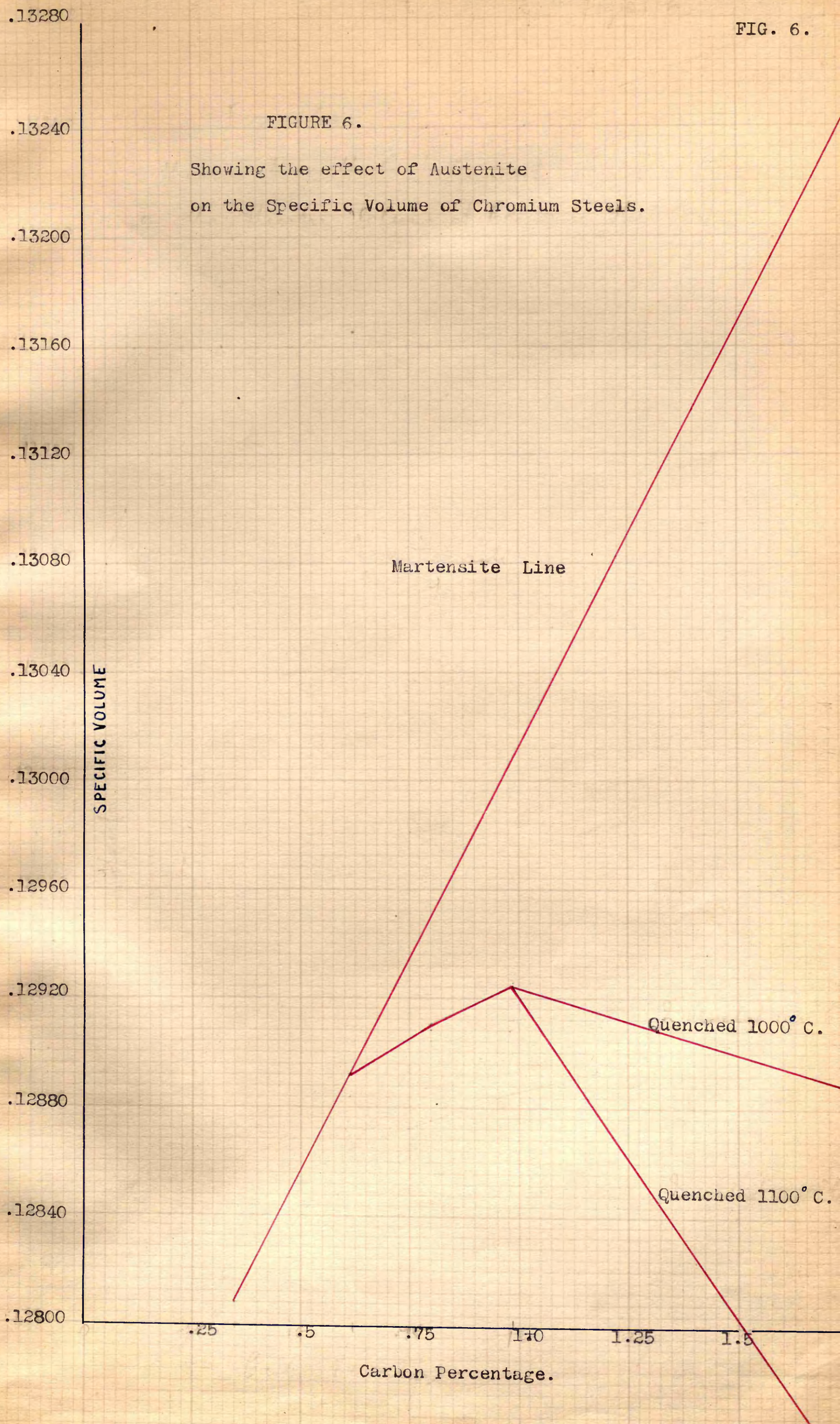


FIGURE 6A.

Showing the percentage of Austenite in carbon steels quenched from 1100° C., assuming the specific volume of martensite to be represented by the Iron-Graphite Mixture line, and the specific volume of austenite to be as calculated from the parameter measurements of Westgren.

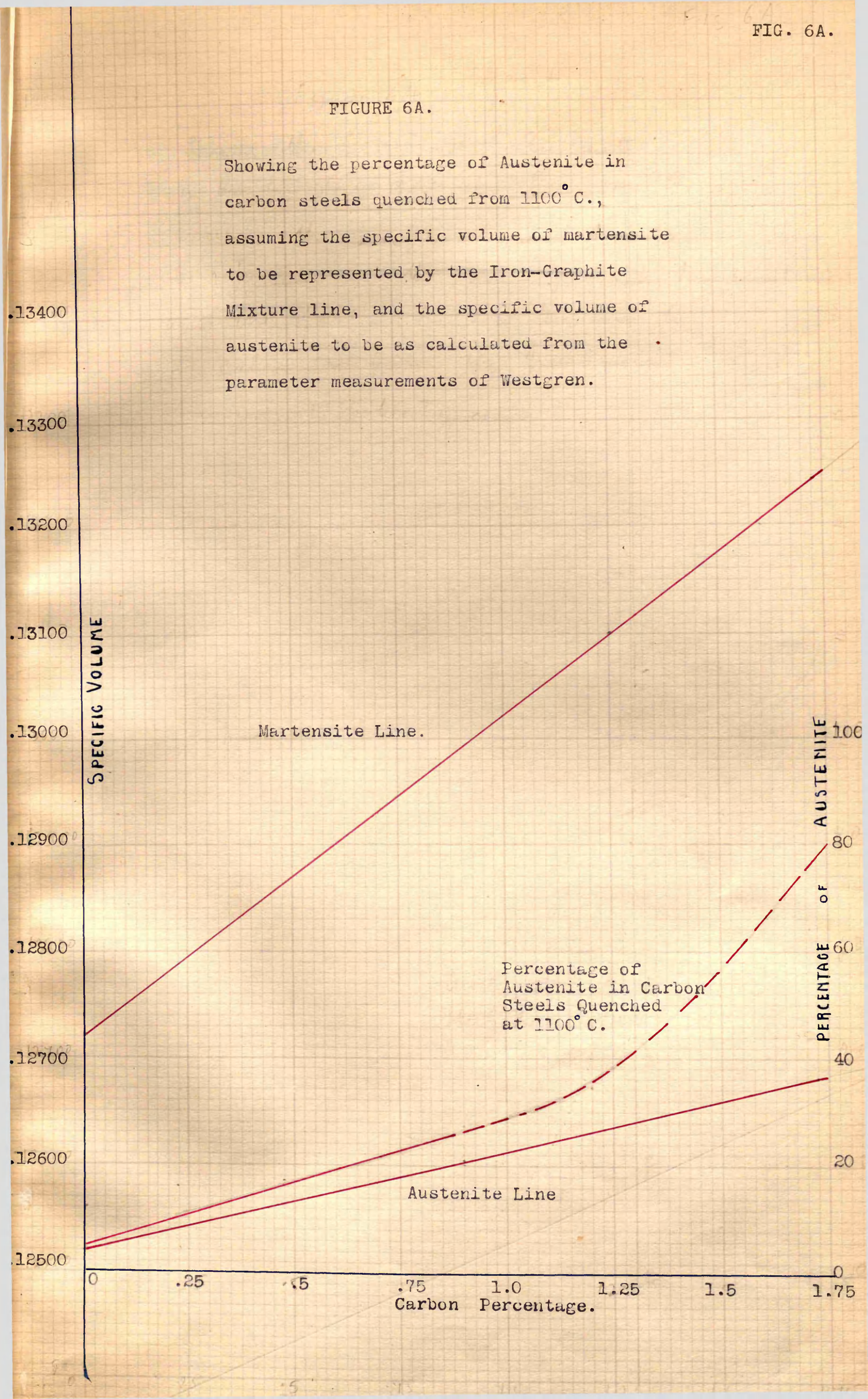


FIG. 7.

FIGURE 7.

The effect of immersion in liquid air on the specific volume of carbon and chromium steels quenched from 1100° C.

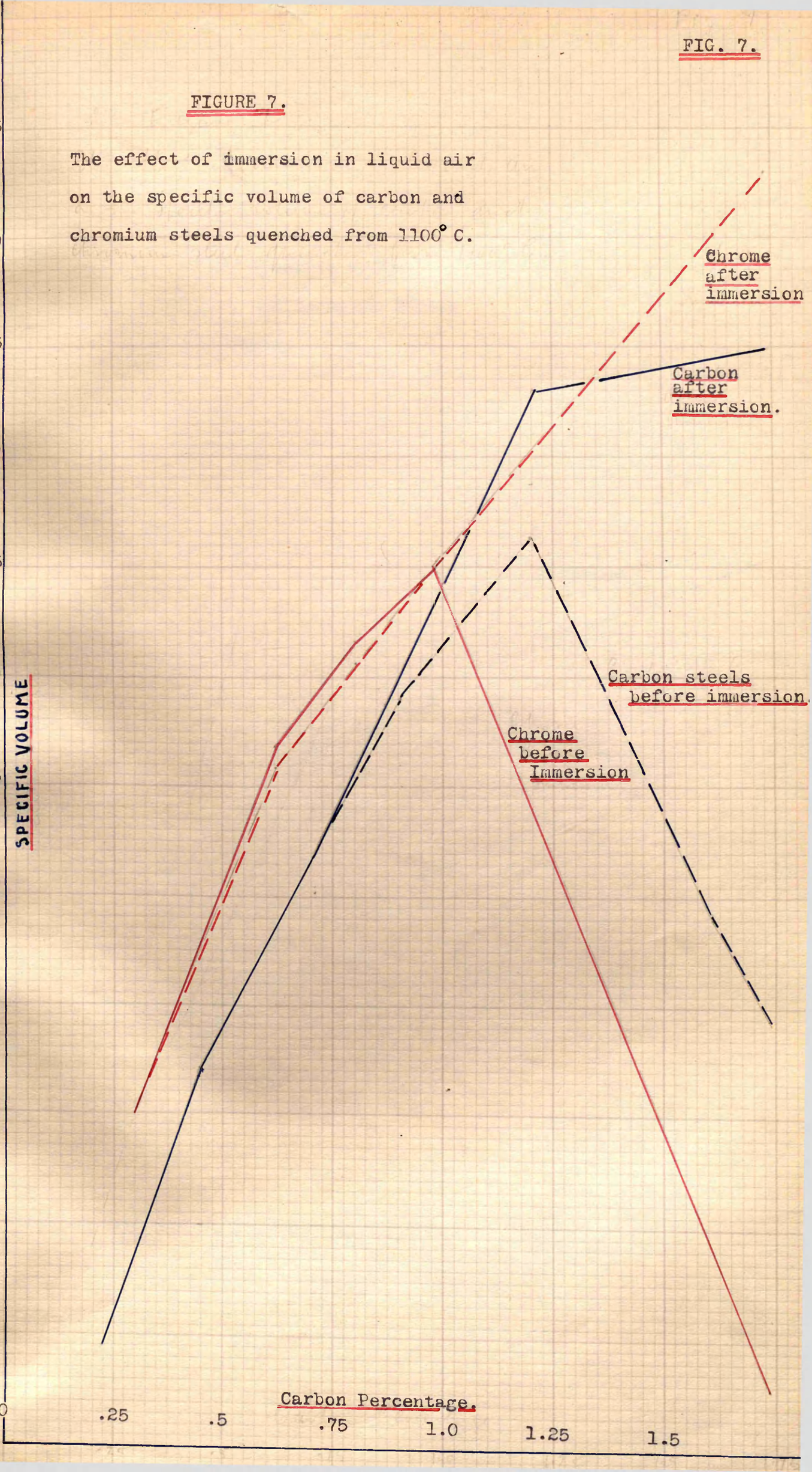


FIGURE 8.

Comparison of the Specific Volume
of Carbon, Nickel and Chromium Steels
in the Annealed State.

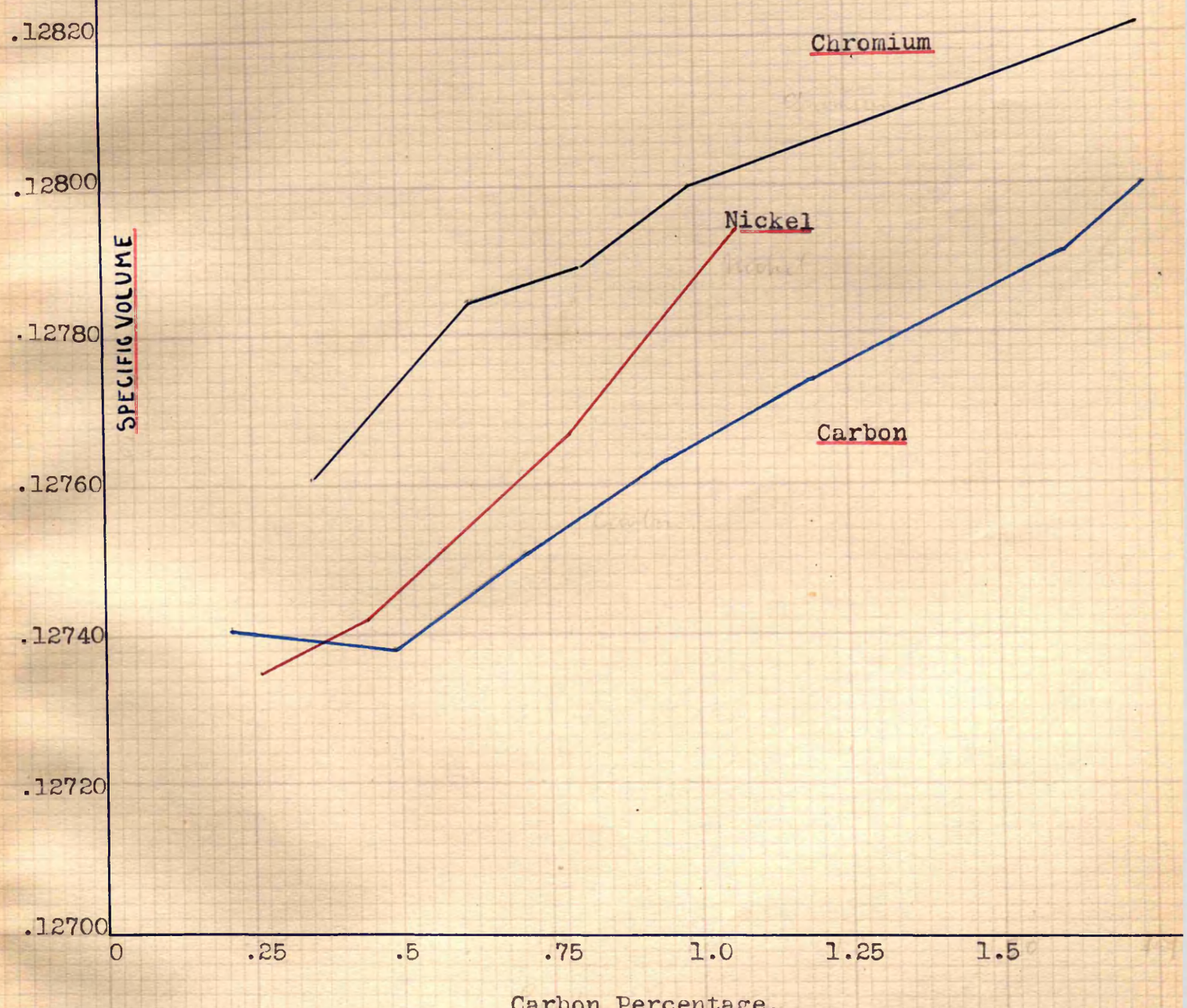


FIGURE 9.

FIG. 9.

Comparison of the Specific Volume
of Carbon, Nickel and Chromium Steels
as Quenched from 1100° C.

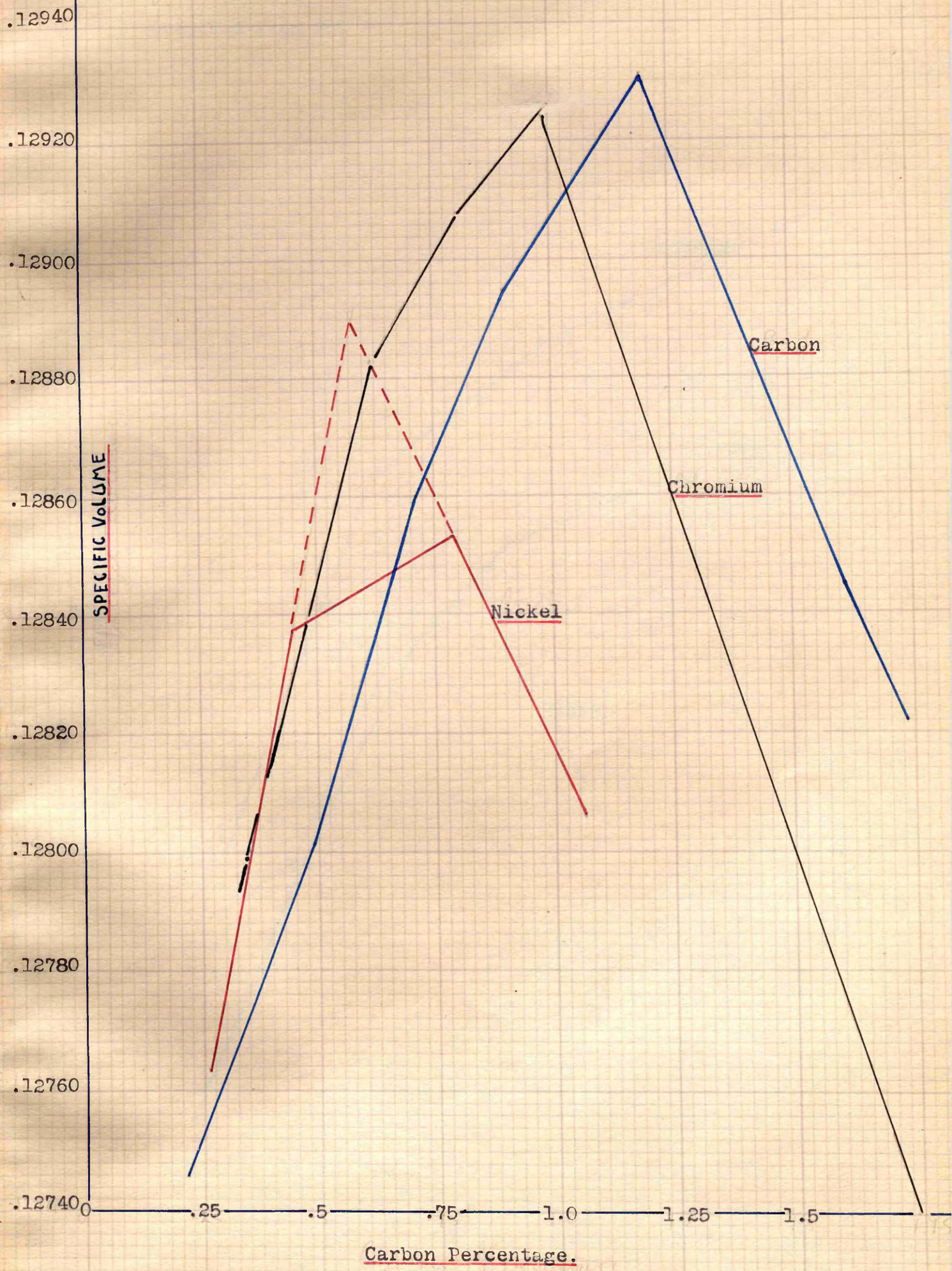
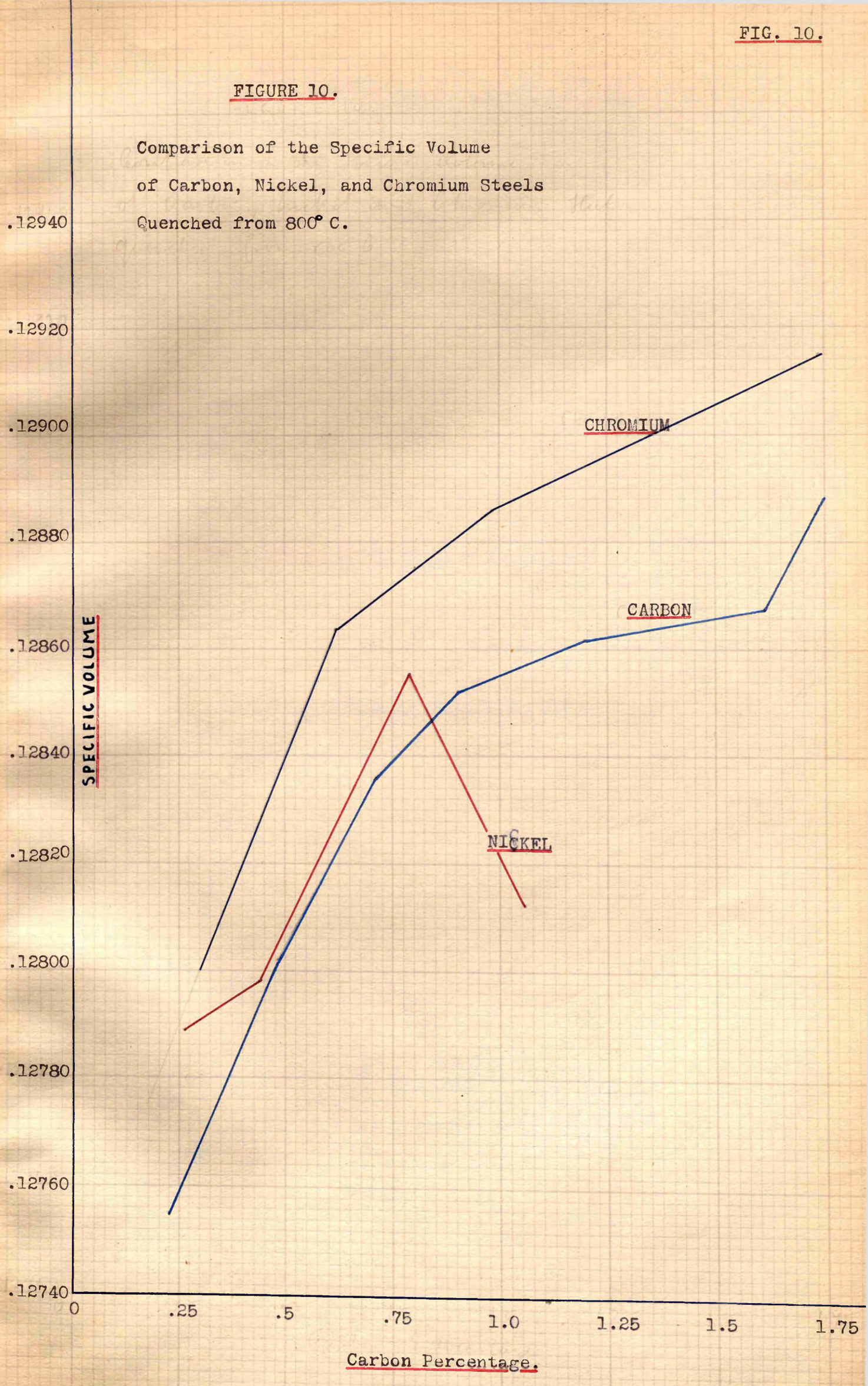
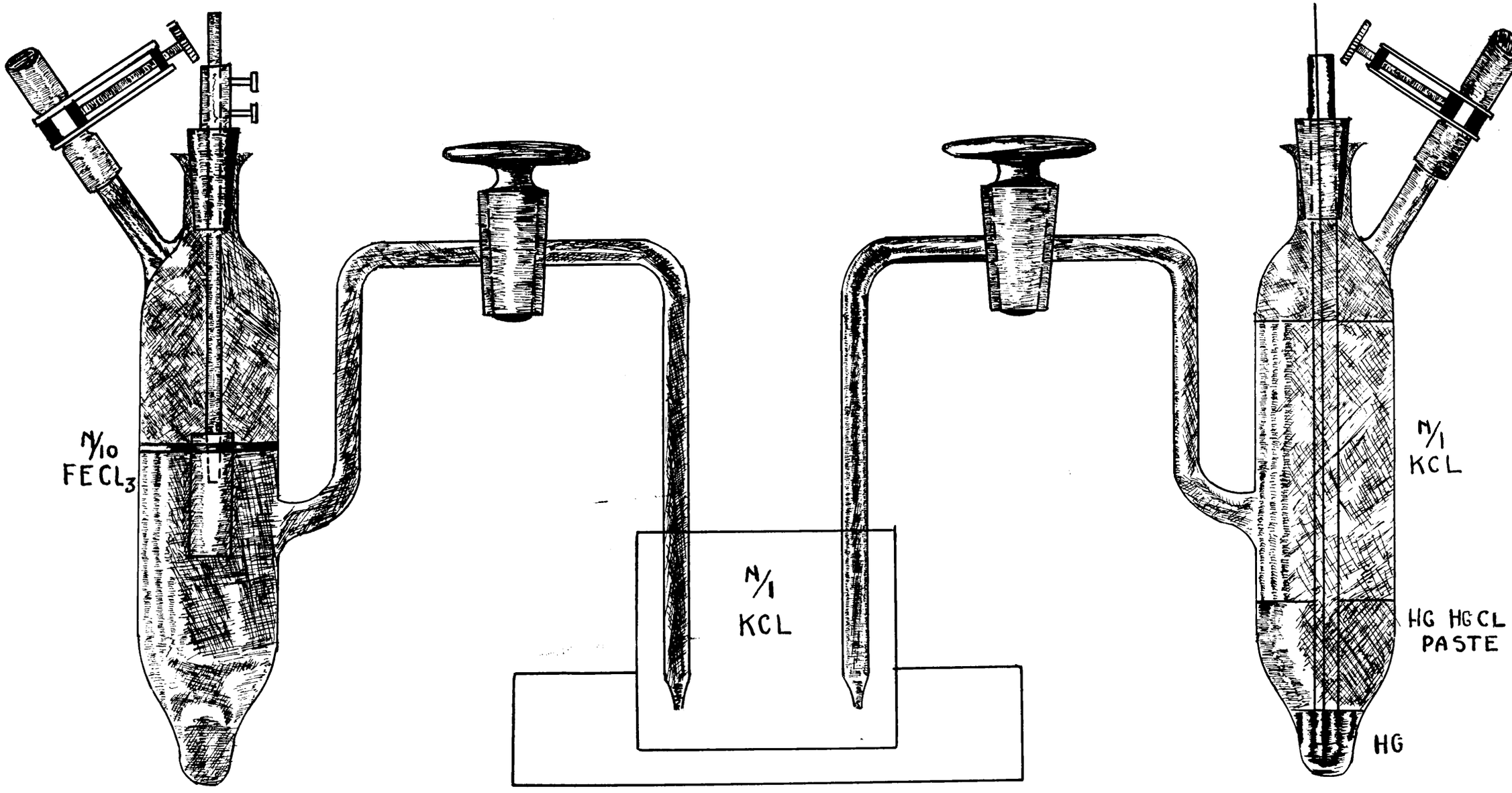


FIGURE 10.

Comparison of the Specific Volume
of Carbon, Nickel, and Chromium Steels
Quenched from 800° C.





APPARATUS
USED
TO
MEASURE
ELECTRODE POTENTIALS

FIG. 11

FIGURE 12 .

Electrode Potential of Carbon and Chromium Steels Quenched from 1000°C.

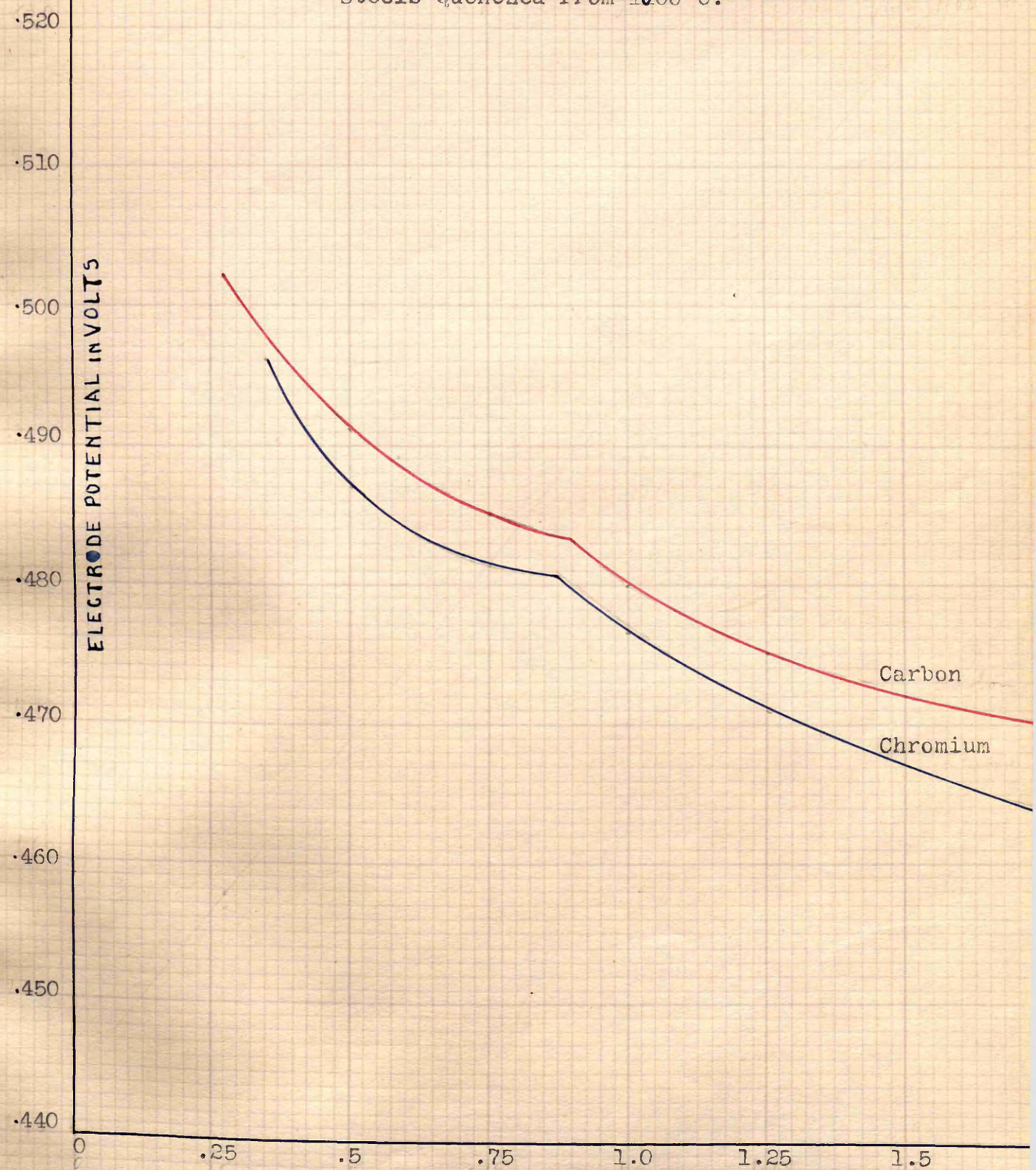


FIGURE 13.

Electrode Potential of Carbon and Chromium Steels Quenched from 1100° C.

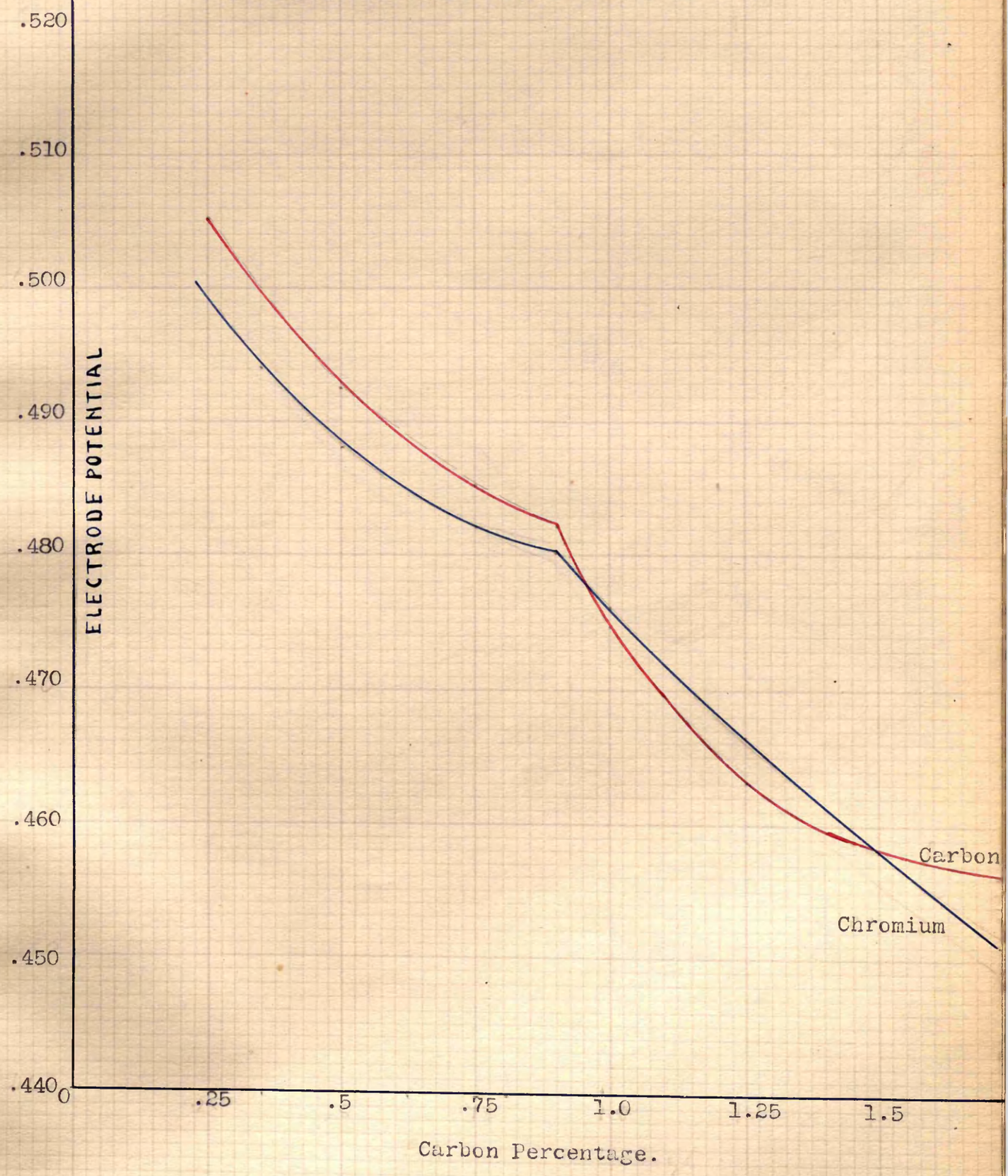


FIGURE 14.

ELECTRODE POTENTIAL of CHROME STEELS
Quenched and Tempered.

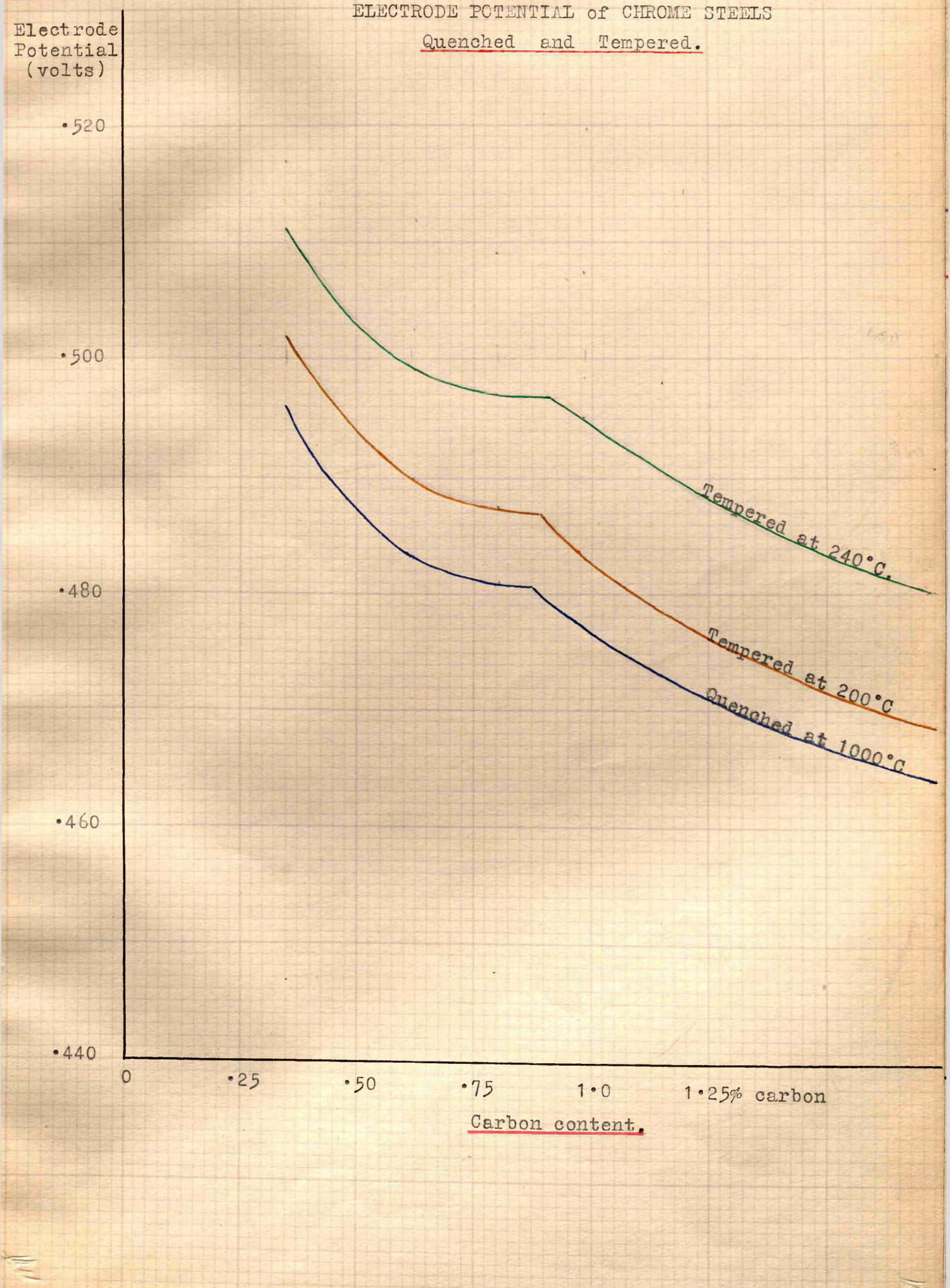


FIG. 14 A.

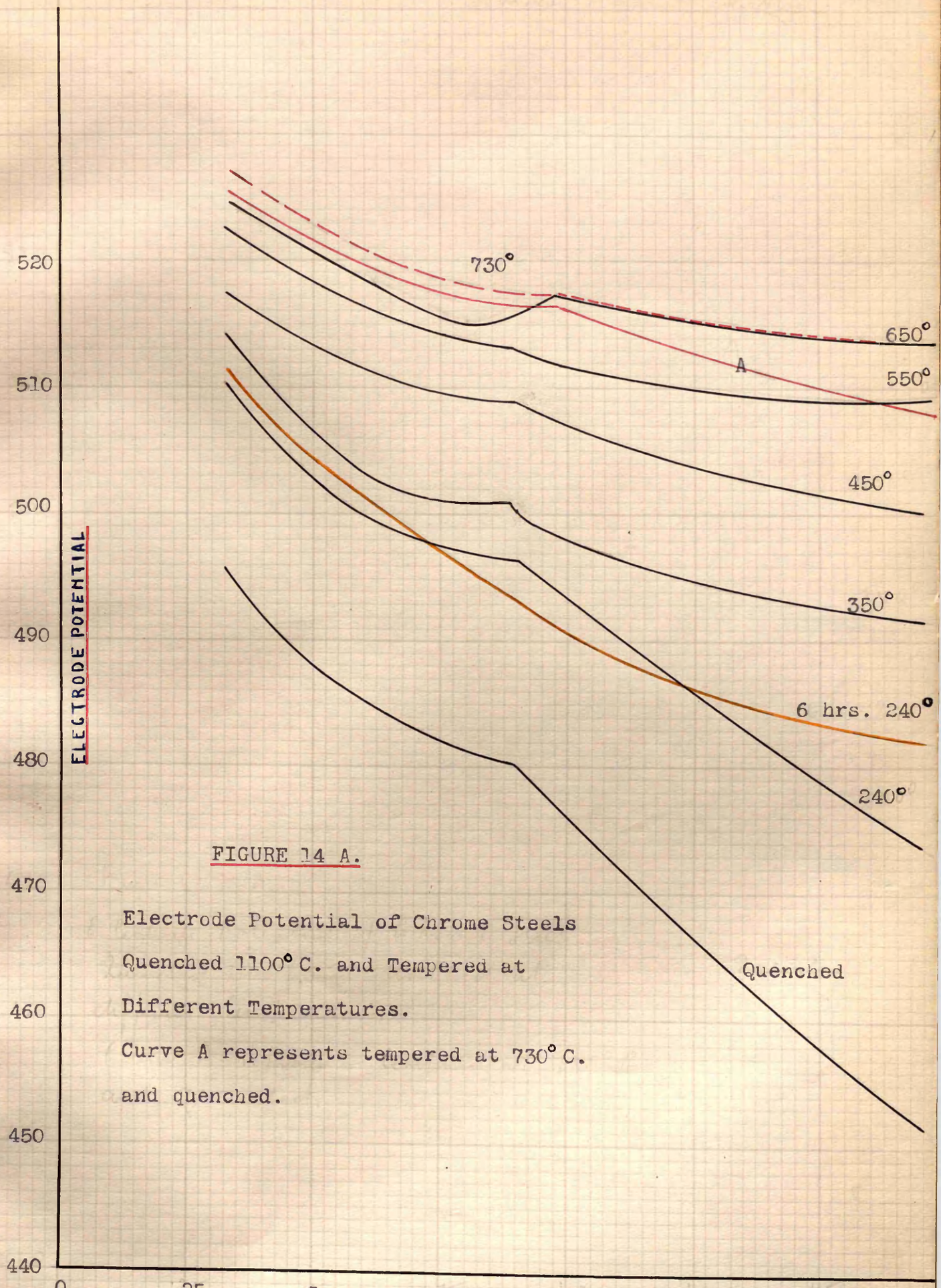


FIGURE 14 A.

Electrode Potential of Chrome Steels Quenched 1100° C. and Tempered at Different Temperatures.

Curve A represents tempered at 730° C. and quenched.

FIGURE 15.

Electrode Potential of Carbon Steels Quenched at 1000°C. and Tempered.

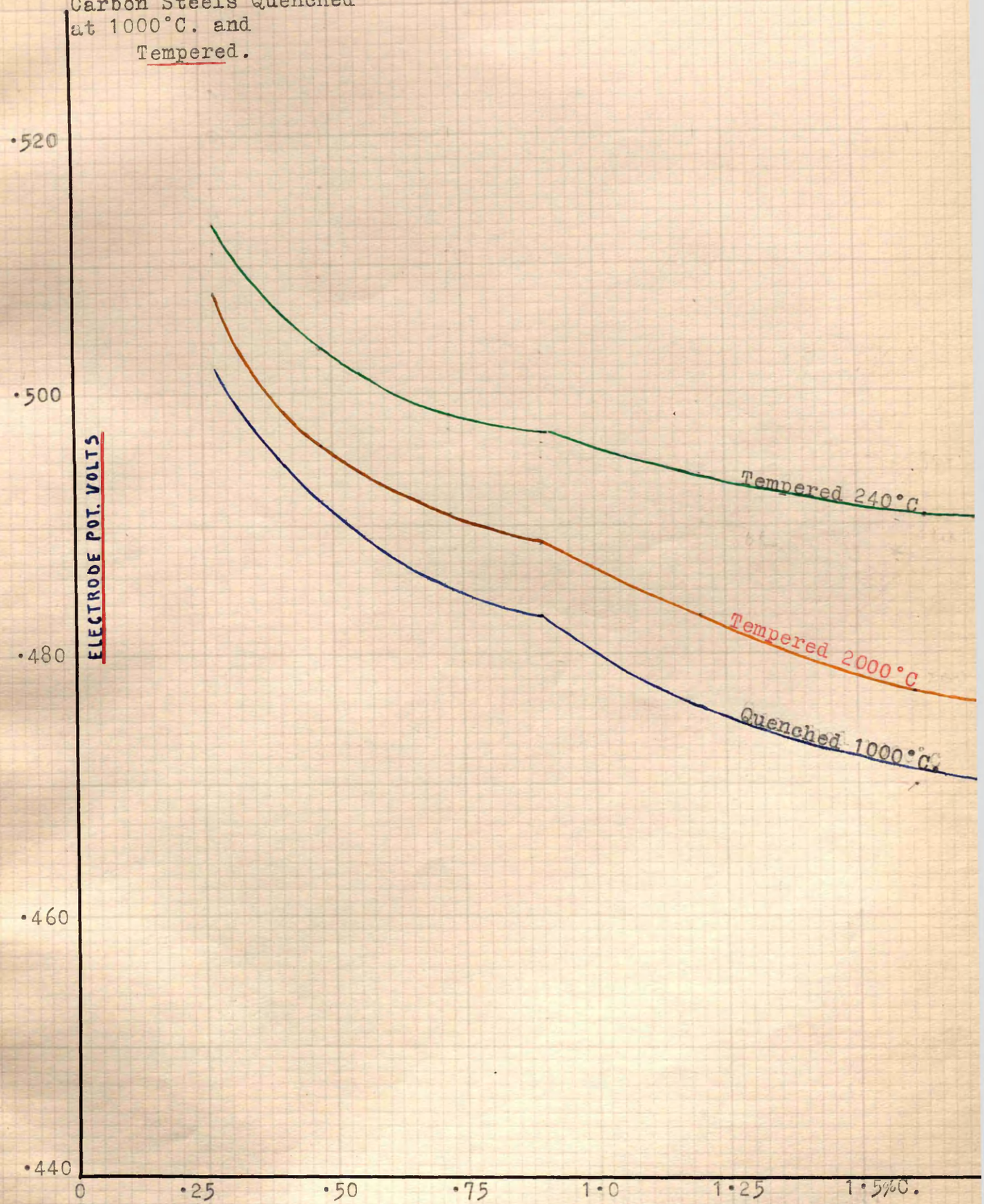


FIG. 16.

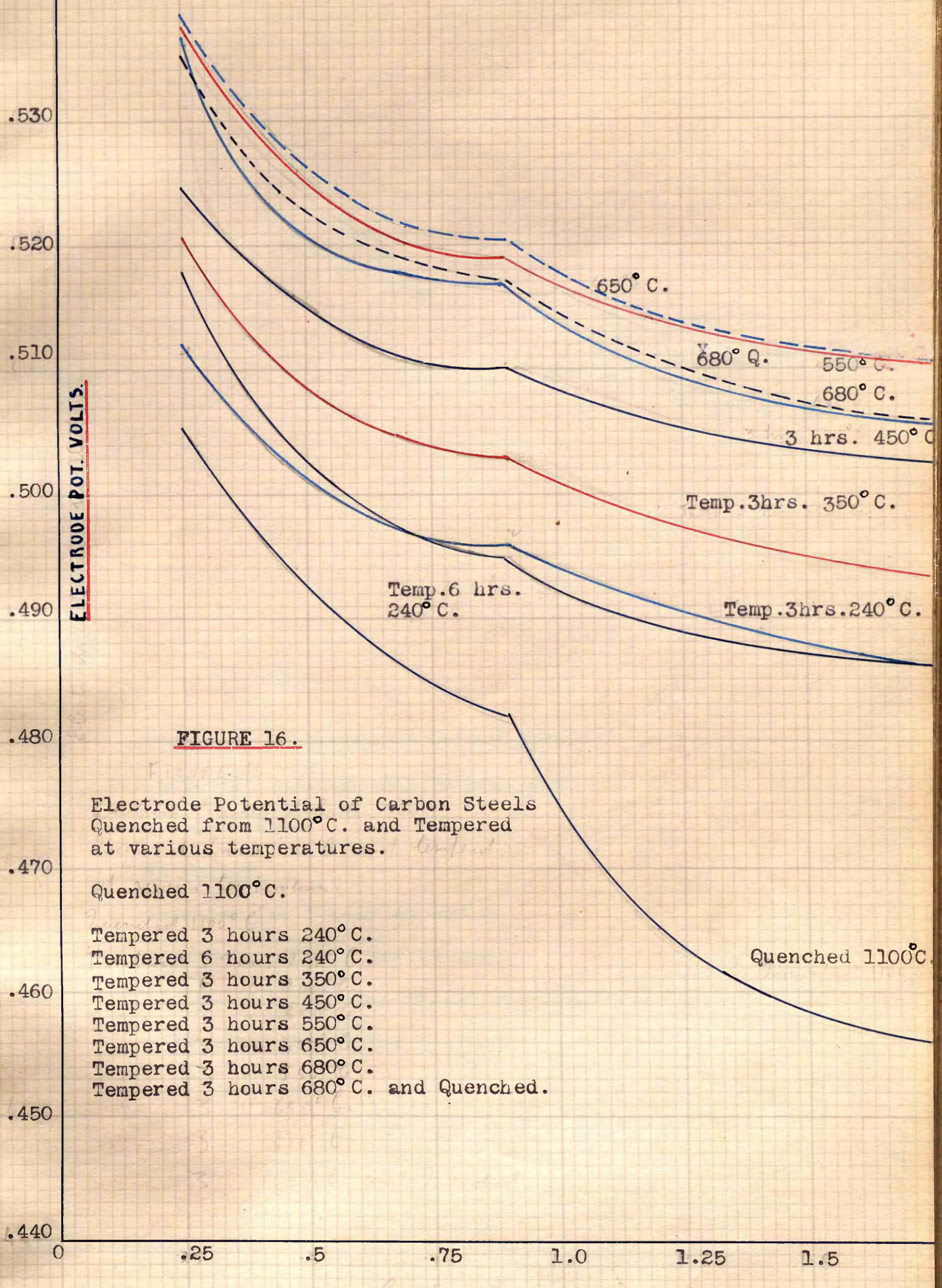


FIGURE 16.

Electrode Potential of Carbon Steels Quenched from 1100° C. and Tempered at various temperatures.

Quenched 1100° C.

Tempered 3 hours 240° C.
Tempered 6 hours 240° C.
Tempered 3 hours 350° C.
Tempered 3 hours 450° C.
Tempered 3 hours 550° C.
Tempered 3 hours 650° C.
Tempered 3 hours 680° C.
Tempered 3 hours 680° C. and Quenched.

FIG. 17.

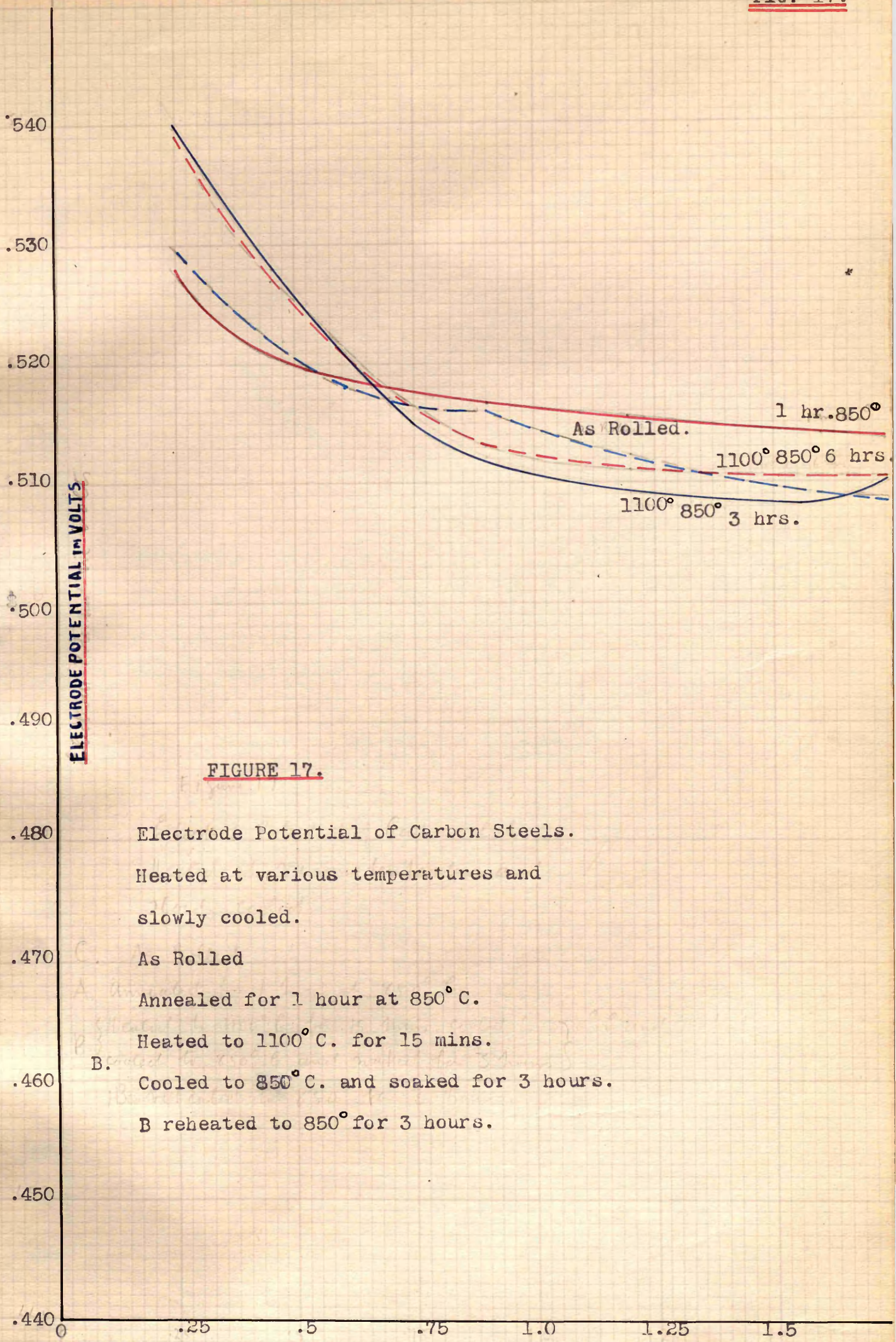


FIGURE 17.

Electrode Potential of Carbon Steels.

Heated at various temperatures and slowly cooled.

As Rolled

Annealed for 1 hour at 850° C.

Heated to 1100° C. for 15 mins.

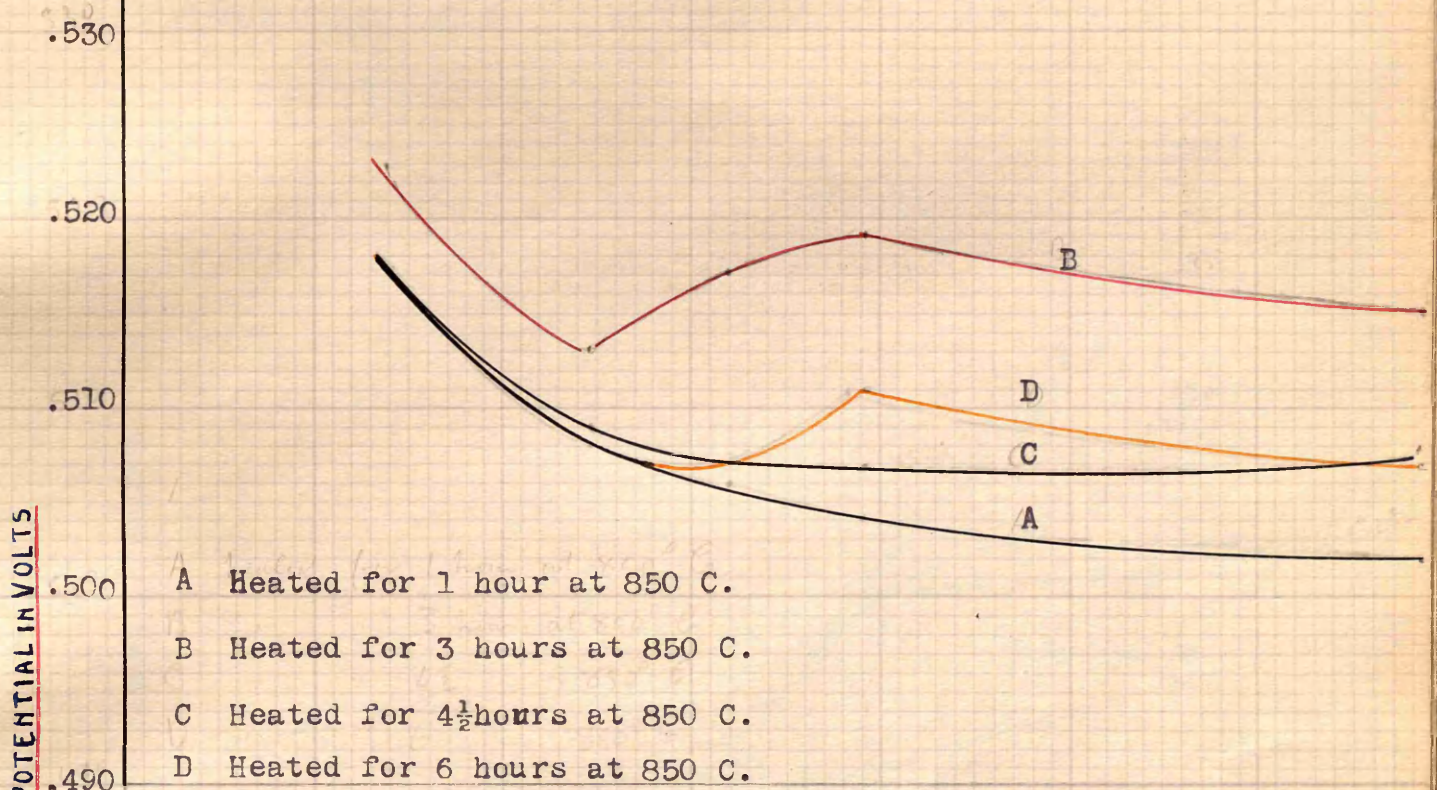
B.

Cooled to 850° C. and soaked for 3 hours.

B reheated to 850° for 3 hours.

FIGURE 18.

Electrode Potential of Chromium Steels.
Heated at Various Temperatures and
Slowly Cooled.

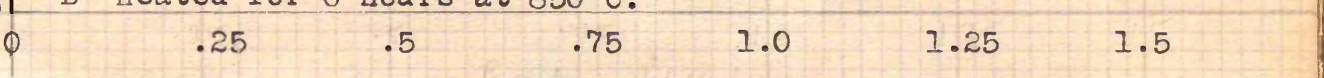


A Heated for 1 hour at 850 C.

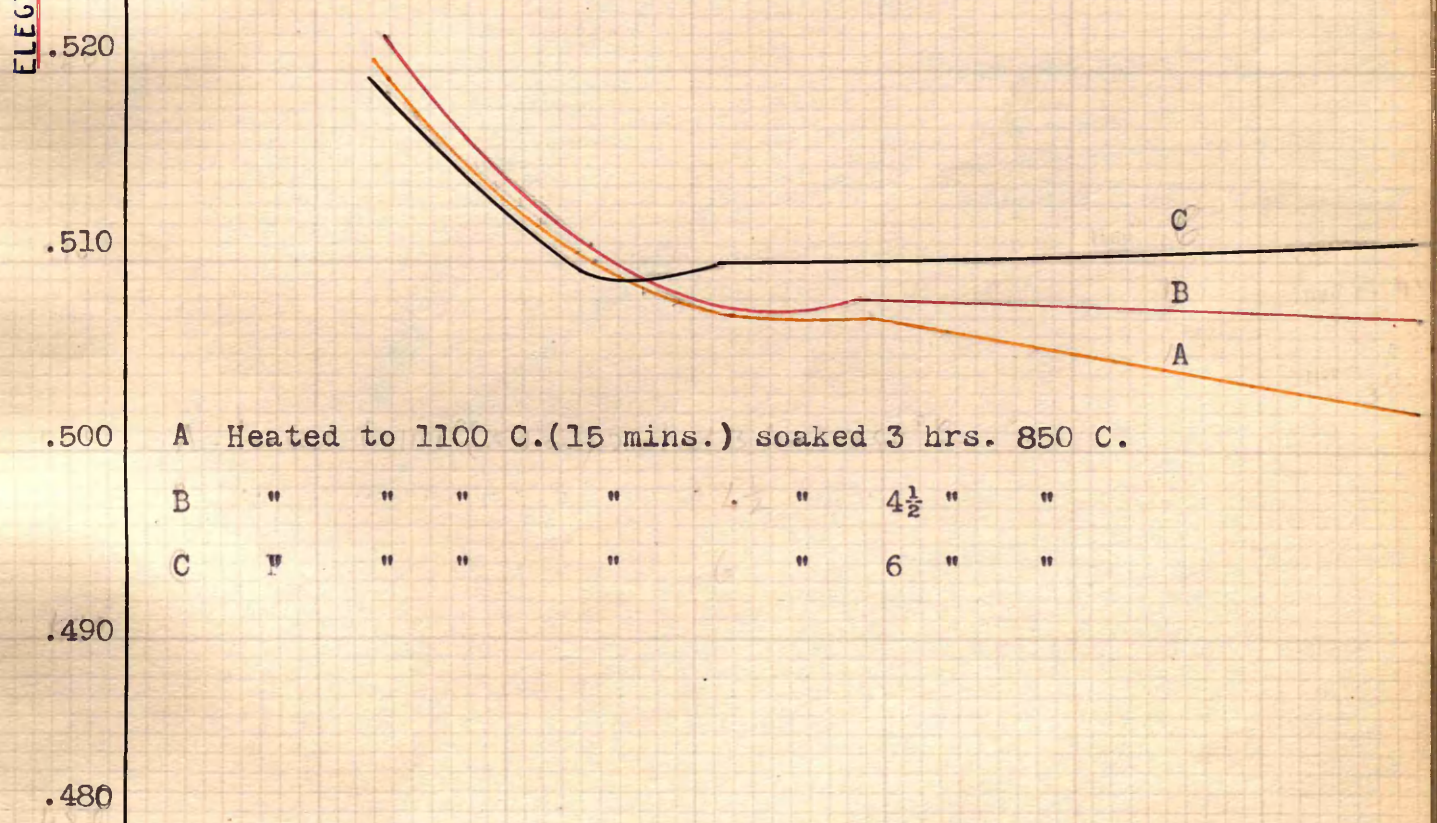
B Heated for 3 hours at 850 C.

C Heated for 4½hours at 850 C.

D Heated for 6 hours at 850 C.



Carbon Percentage.



A Heated to 1100 C.(15 mins.) soaked 3 hrs. 850 C.

B " " " " " 4½ " "

C " " " " " 6 " "

Carbon Percentage.

FIGURE 19.

Electrical Resistance of Chromium Steels.

Quenched from Different Temperatures.

Tempered at 240° C.

And as Annealed.

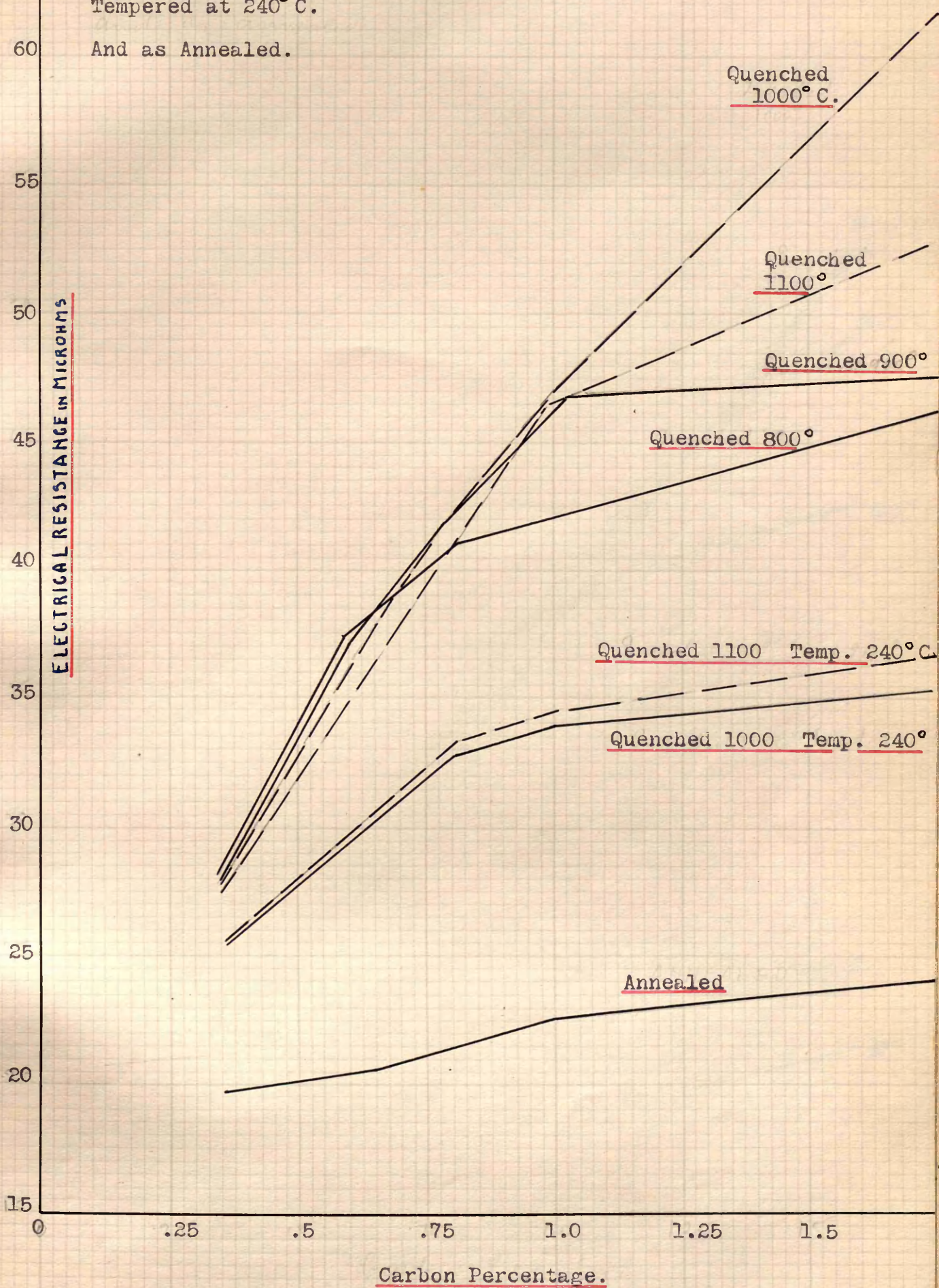


FIGURE 20.

Electrical Resistance of Carbon Steels.

Quenched from Different Temperatures.

Tempered at 240°C. and Annealed.

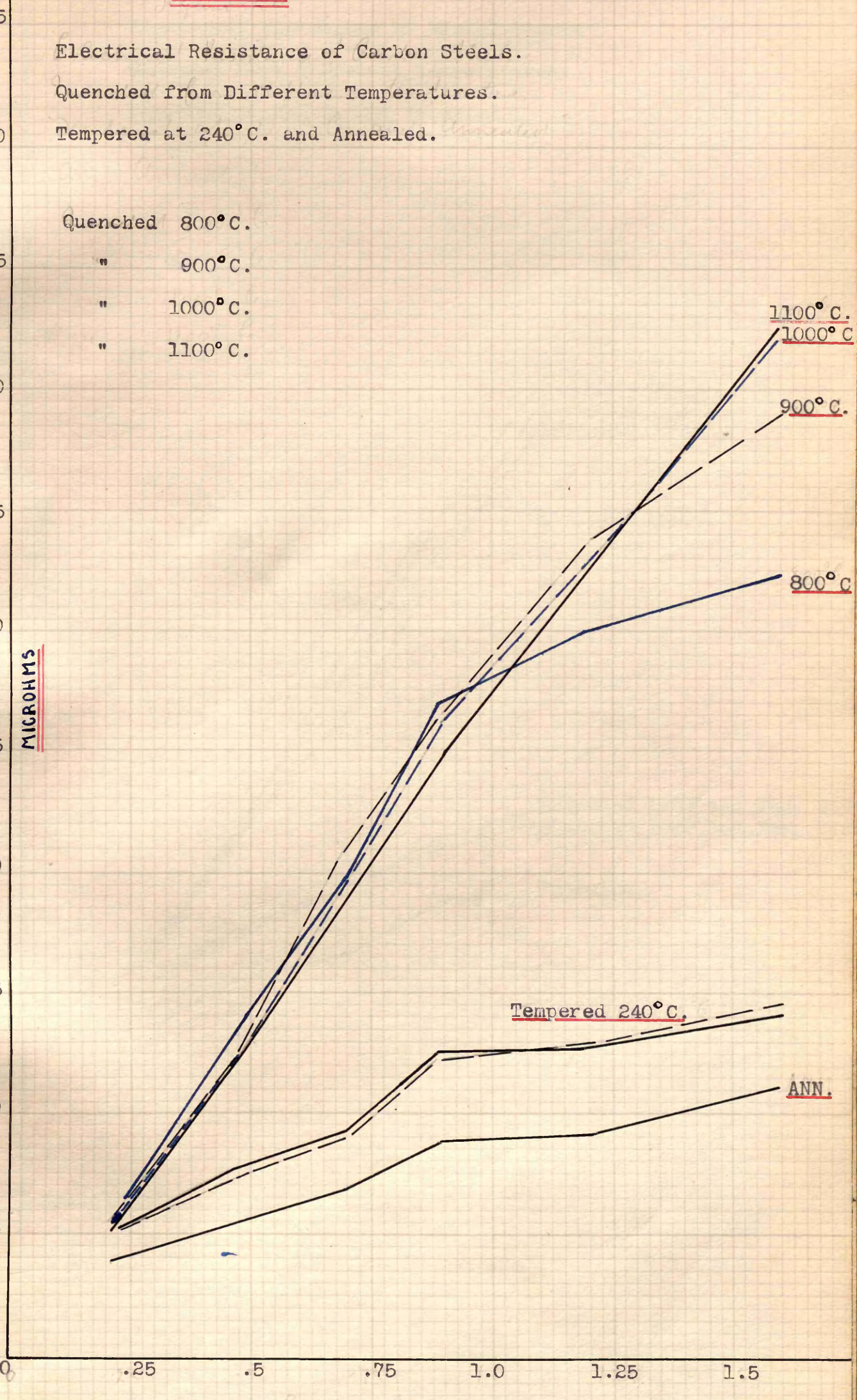


FIGURE 21.

Electrical Resistance of Nickel Steels
Quenched from Different Temperatures.
Tempered at 240° C. and Annealed.

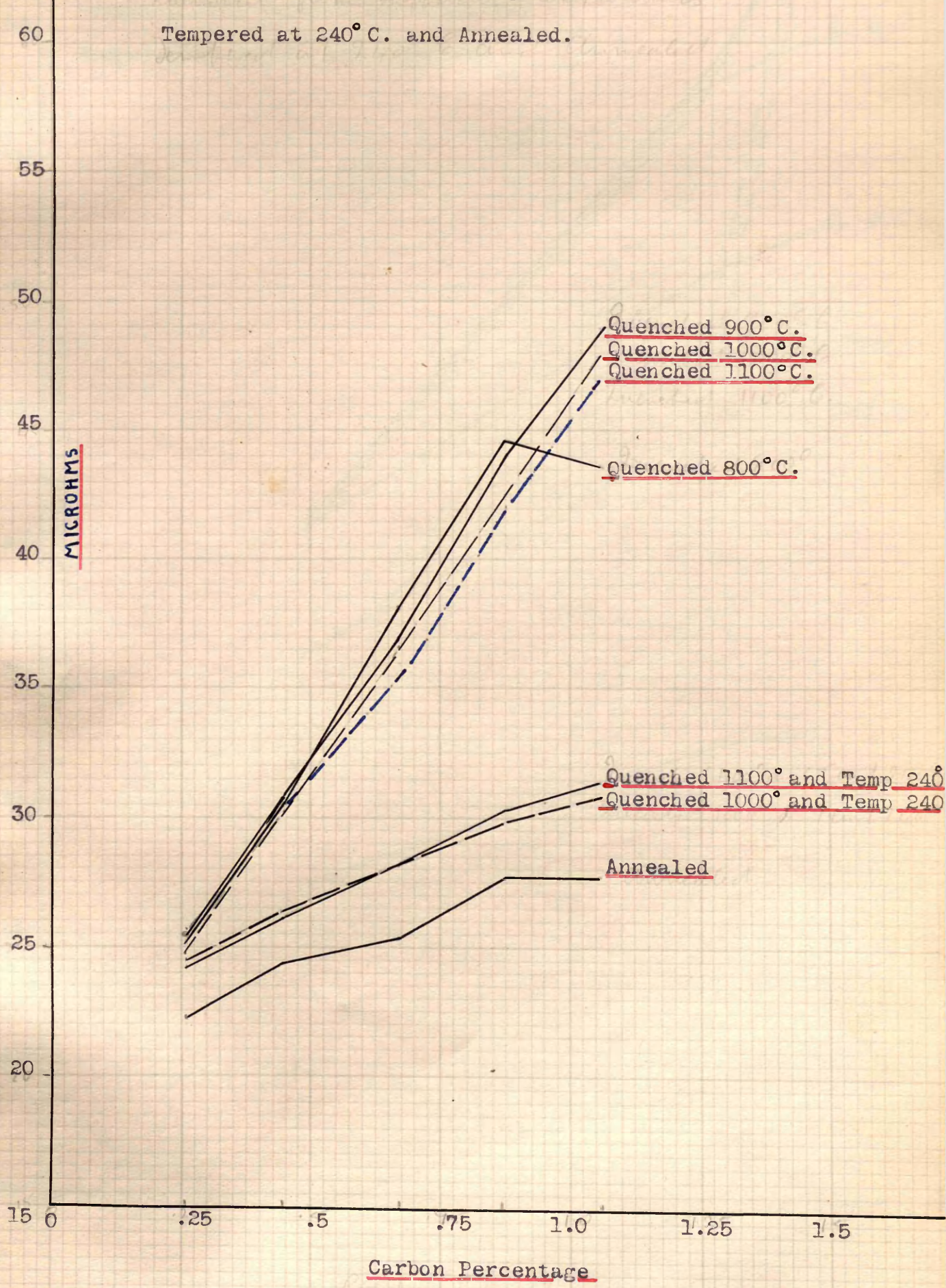


FIG. 22

FIGURE 22.

Electrical Resistance of Nickel-Chrome Steels
Quenched from Various Temperatures.
Tempered at 240° C. and Annealed.

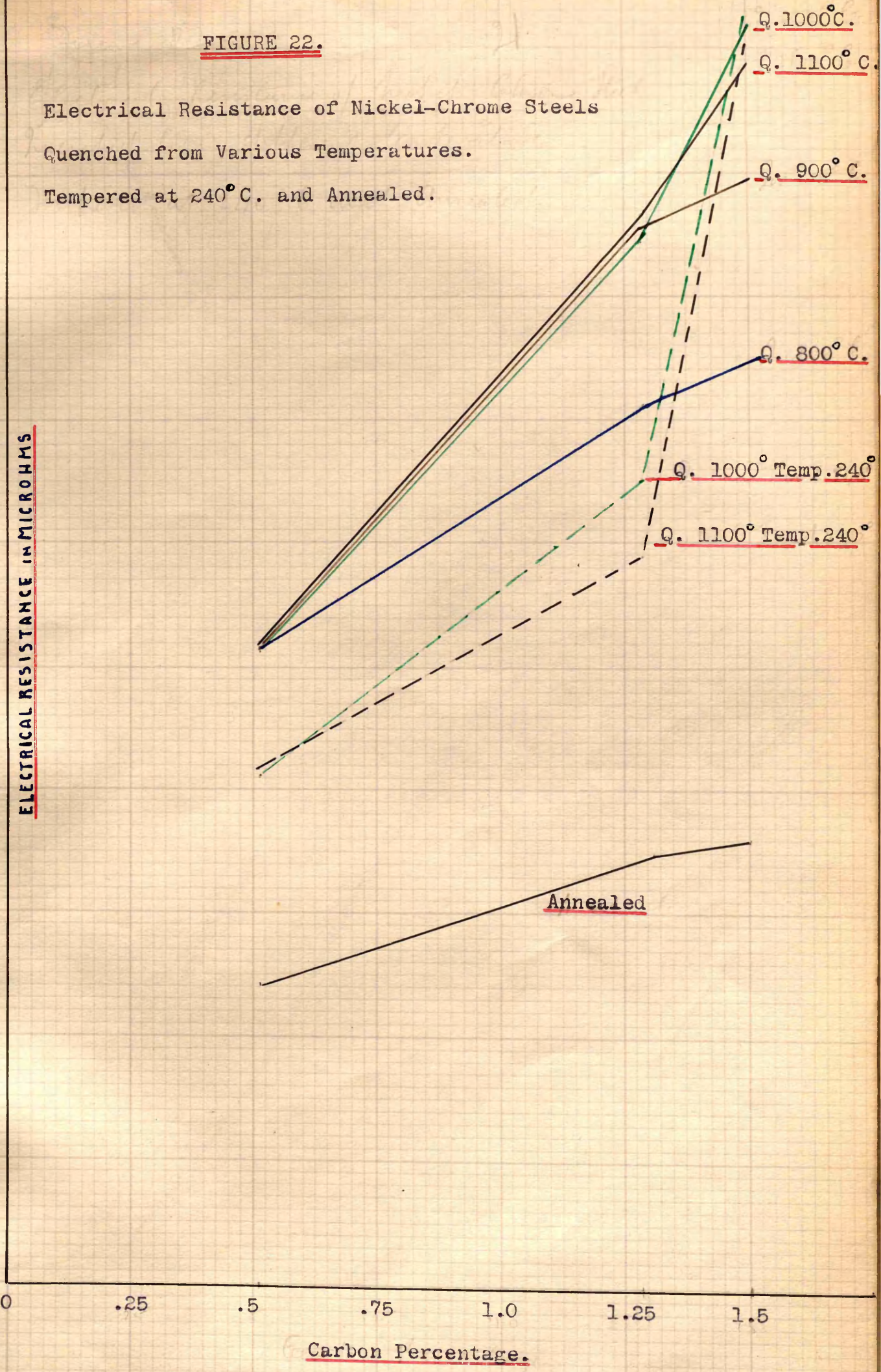


FIGURE 23.

Comparison of the Electrical Resistance of Carbon, Nickel, Chromium and Nickel-Chromium Steels Quenched from 1100° C.

Readings are corrected for Silicon, Phosphoros and Manganese.

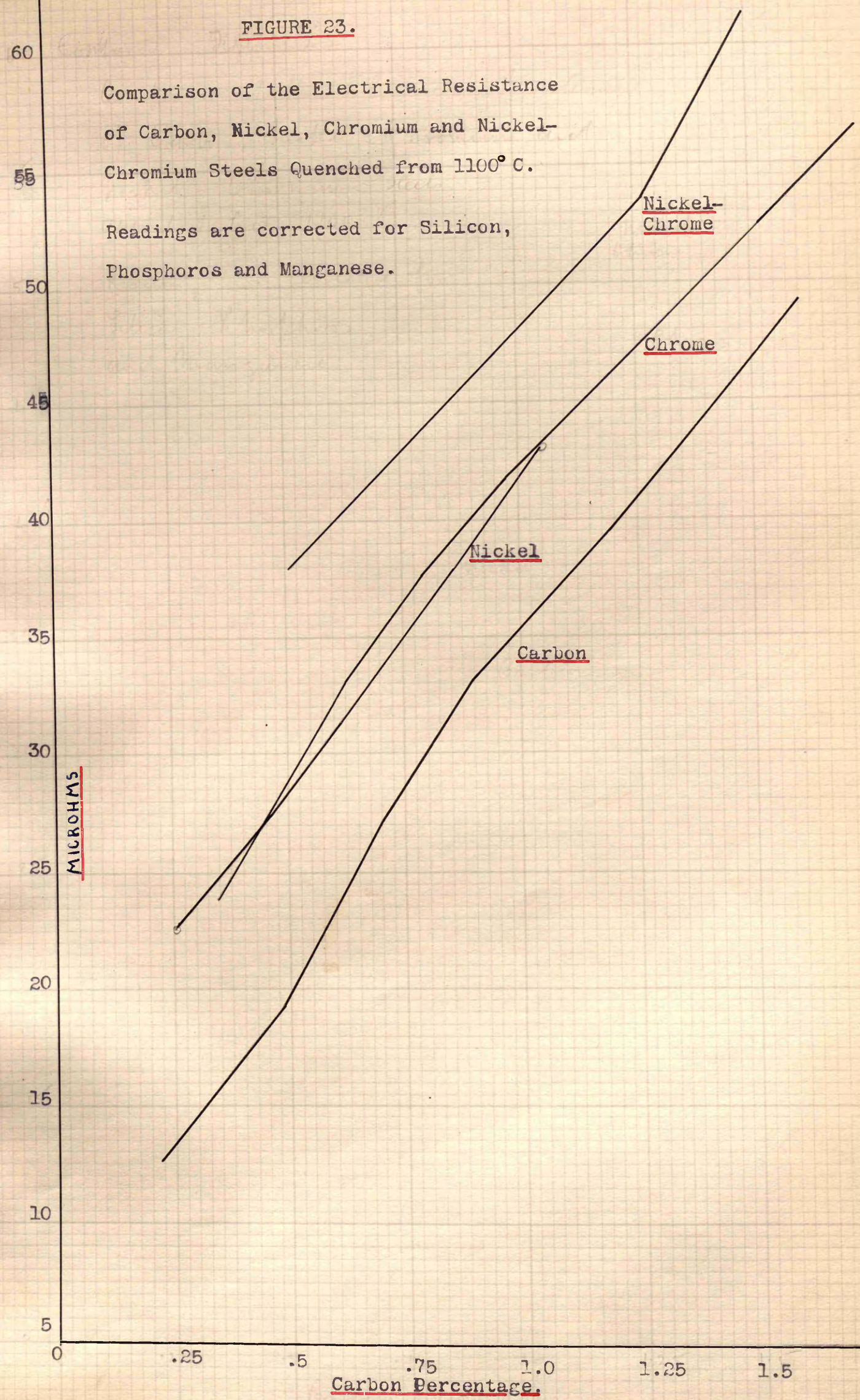


FIGURE 24.

Comparison of the Electrical Resistance of Carbon, Nickel, Chromium, and Nickel-Chromium Steels in Annealed State.

Readings are corrected for Silicon, Phosphoros and Manganese.

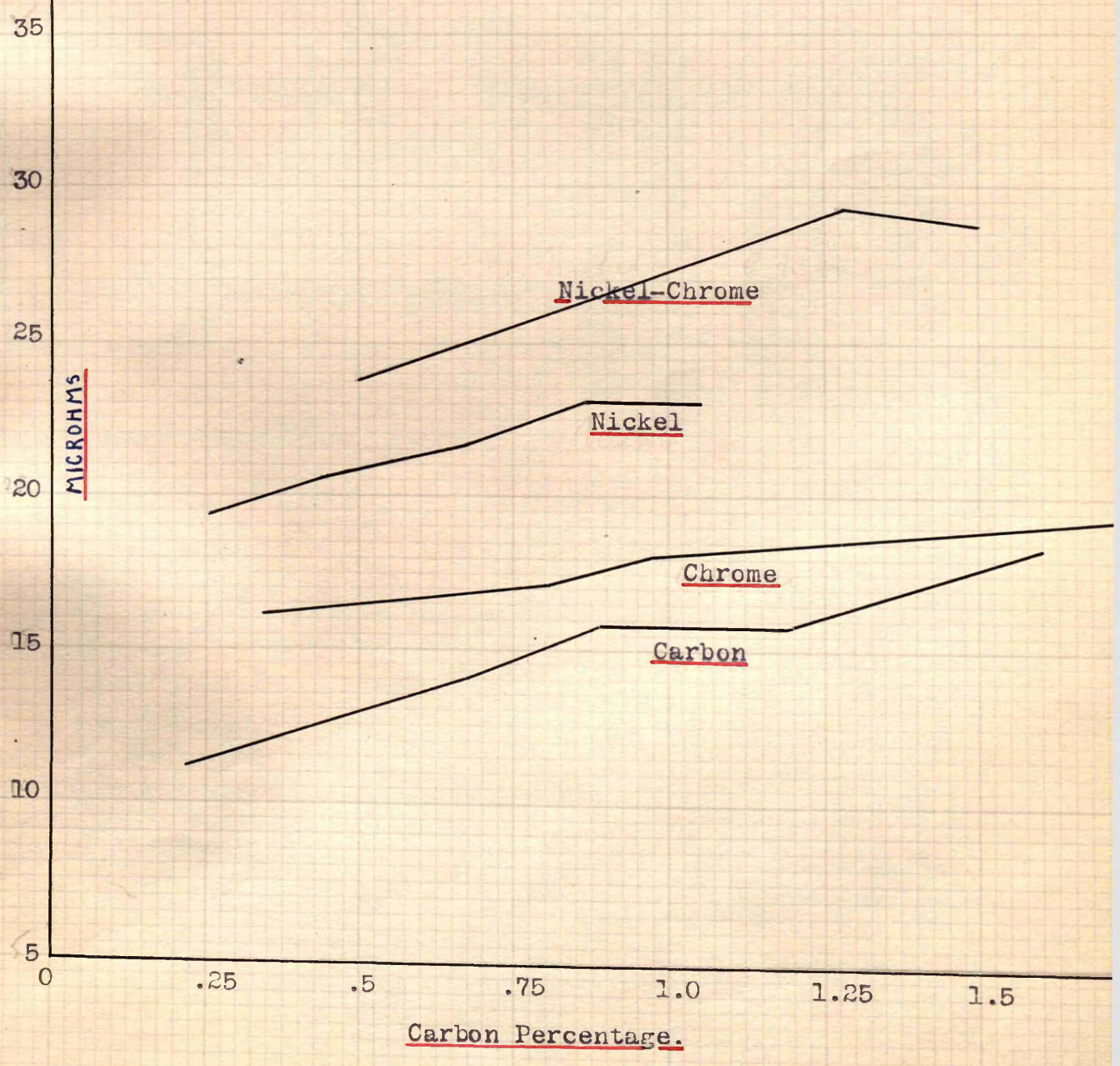
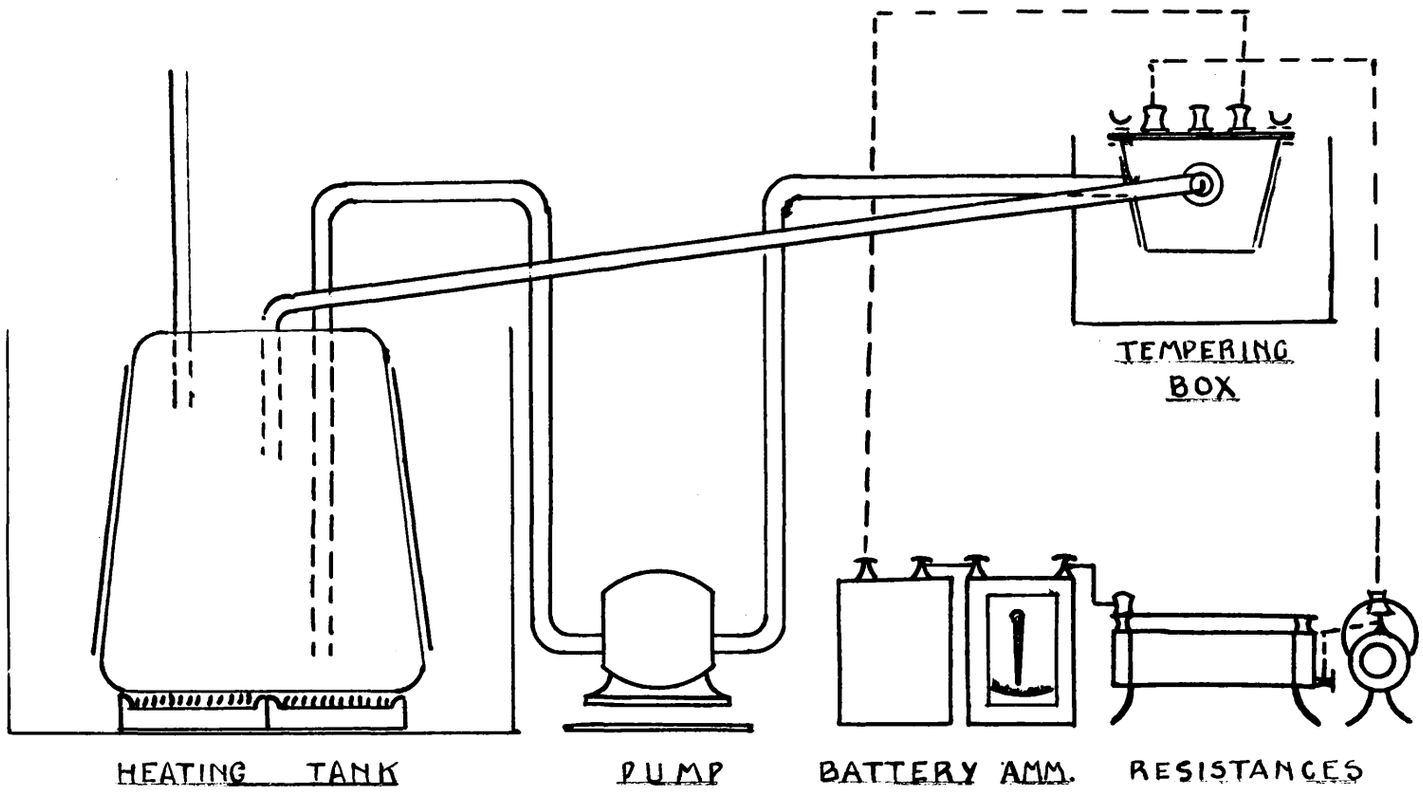
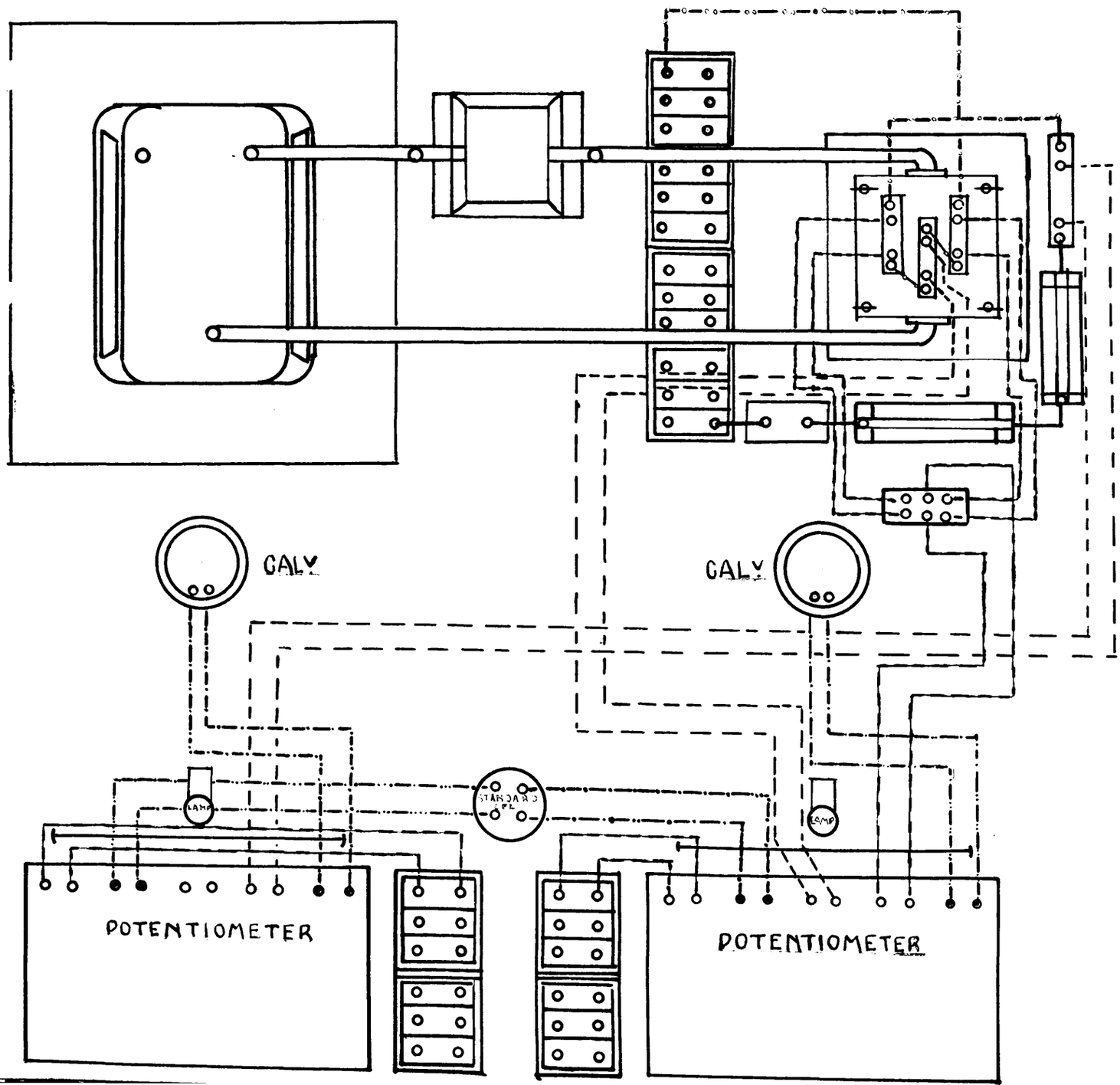
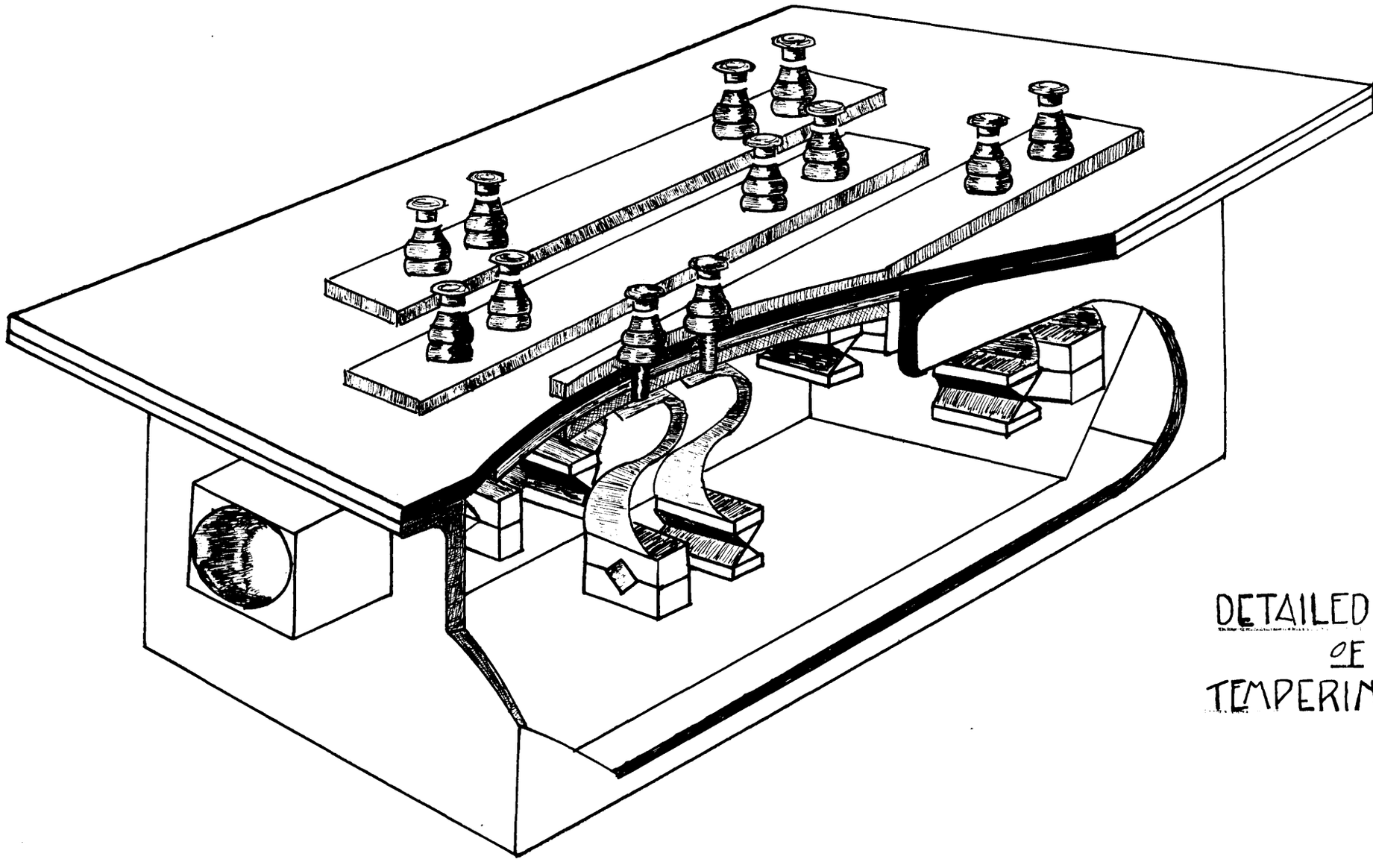


FIG. 25


TEMPERING BOX
HEATING TANK
PUMP
BATTERY AMM.
RESISTANCES


CALV
CALV
LAMP
STANDARD CELL
LAMP
POTENTIOMETER
POTENTIOMETER



DETAILED VIEW
OF
TEMPERING BOX

FIG. 26

FIGURE 27.

TEMPERING-RESISTANCE CURVES OF CARBON STEELS.

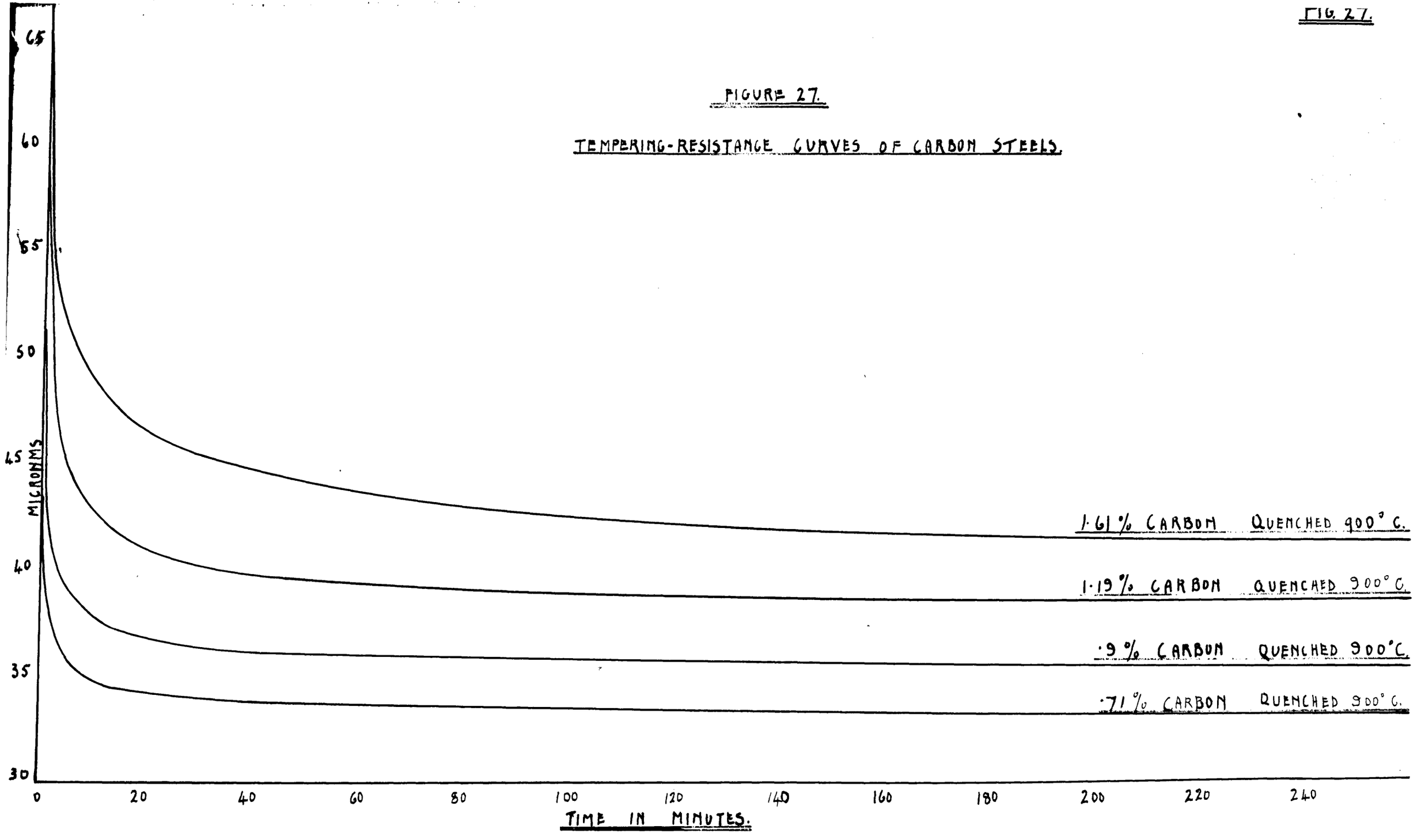


FIGURE 28.

TEMPERING- RESISTANCE CURVES OF CARBON STEELS.

RED POINTS ARE CALCULATED FROM FORMULA.

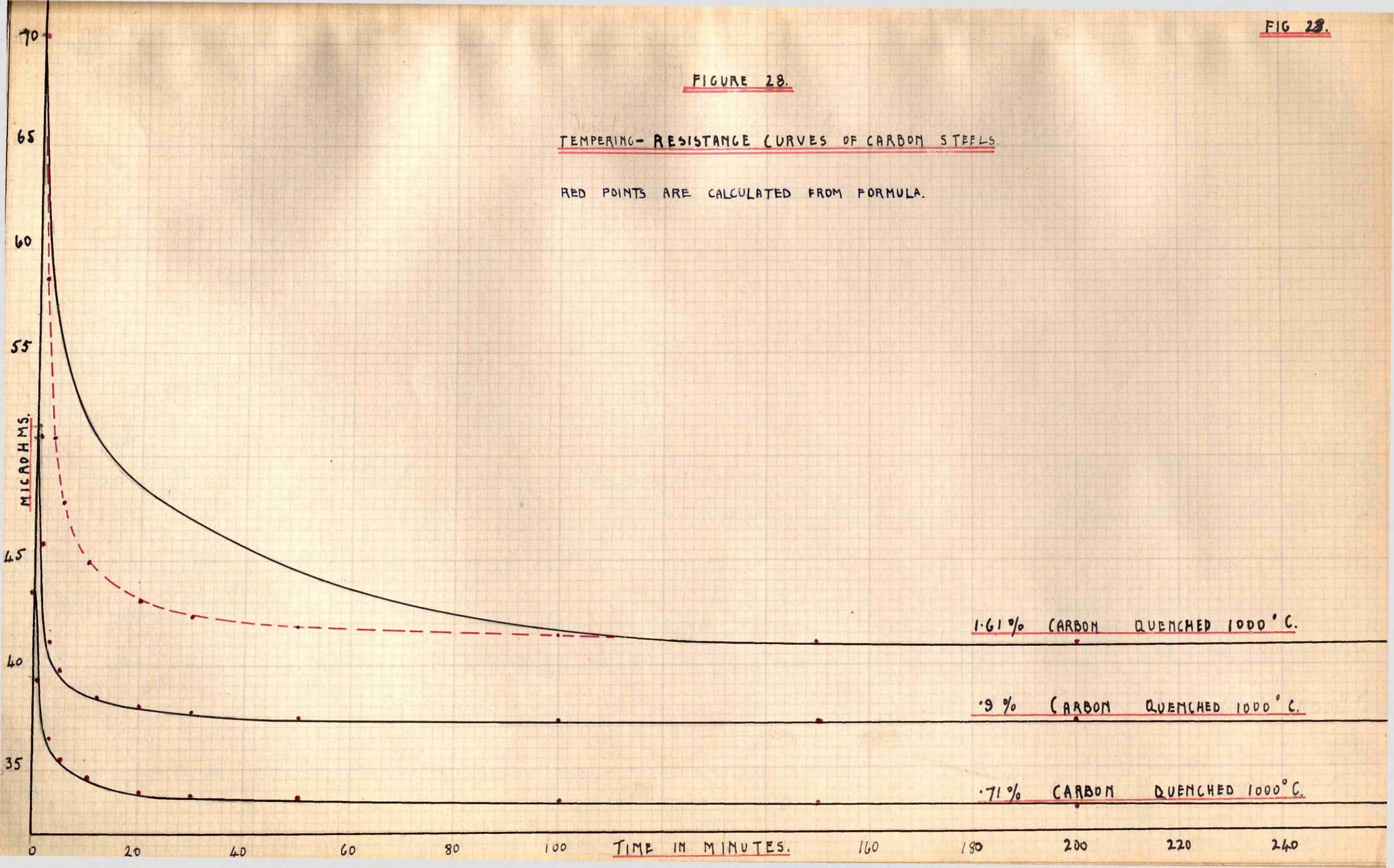


FIGURE 29.

TEMPERING - RESISTANCE CURVES OF CARBON STEELS.

RED POINTS ARE CALCULATED FROM FORMULA.

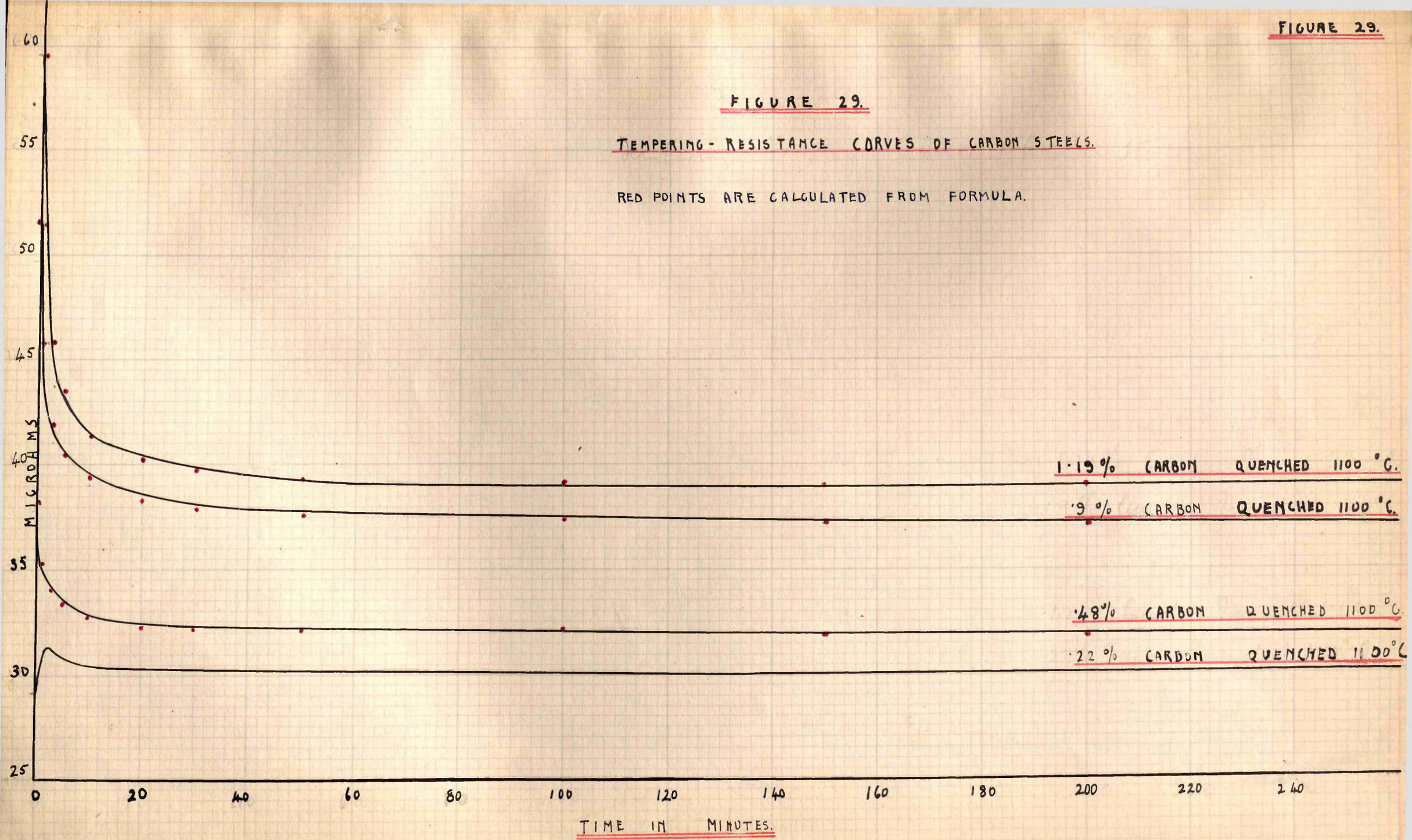


FIGURE 30.

TEMPERING - RESISTANCE CURVES OF CARBON STEELS.

RED POINTS ARE CALCULATED FROM FORMULA.

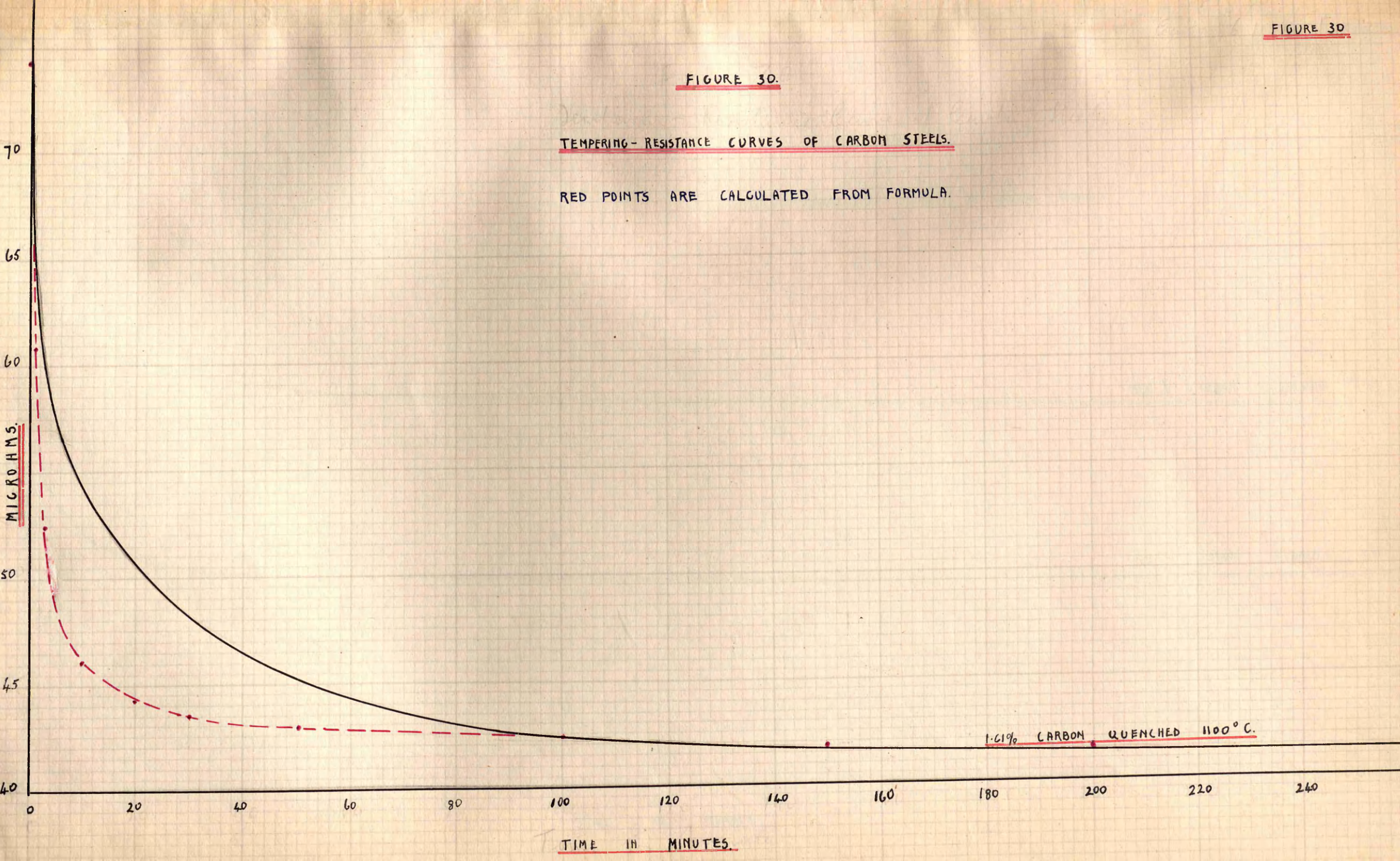


FIG. 31.

FIGURE 31.

TEMPERING-RESISTANCE CURVES OF CHROME STEELS.

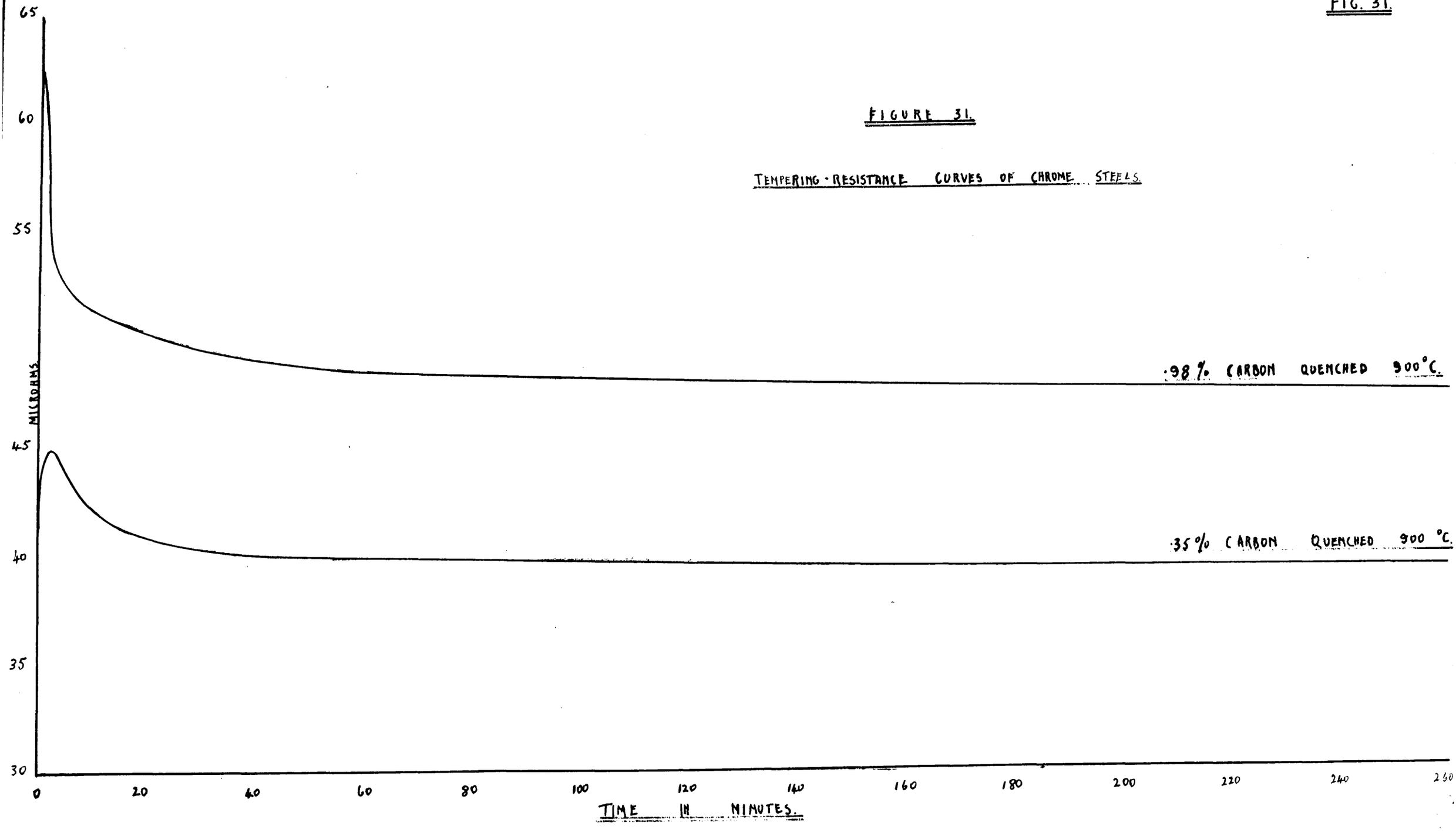


FIGURE 32.

TEMPERING – RESISTANCE CURVES OF CHROME STEELS.

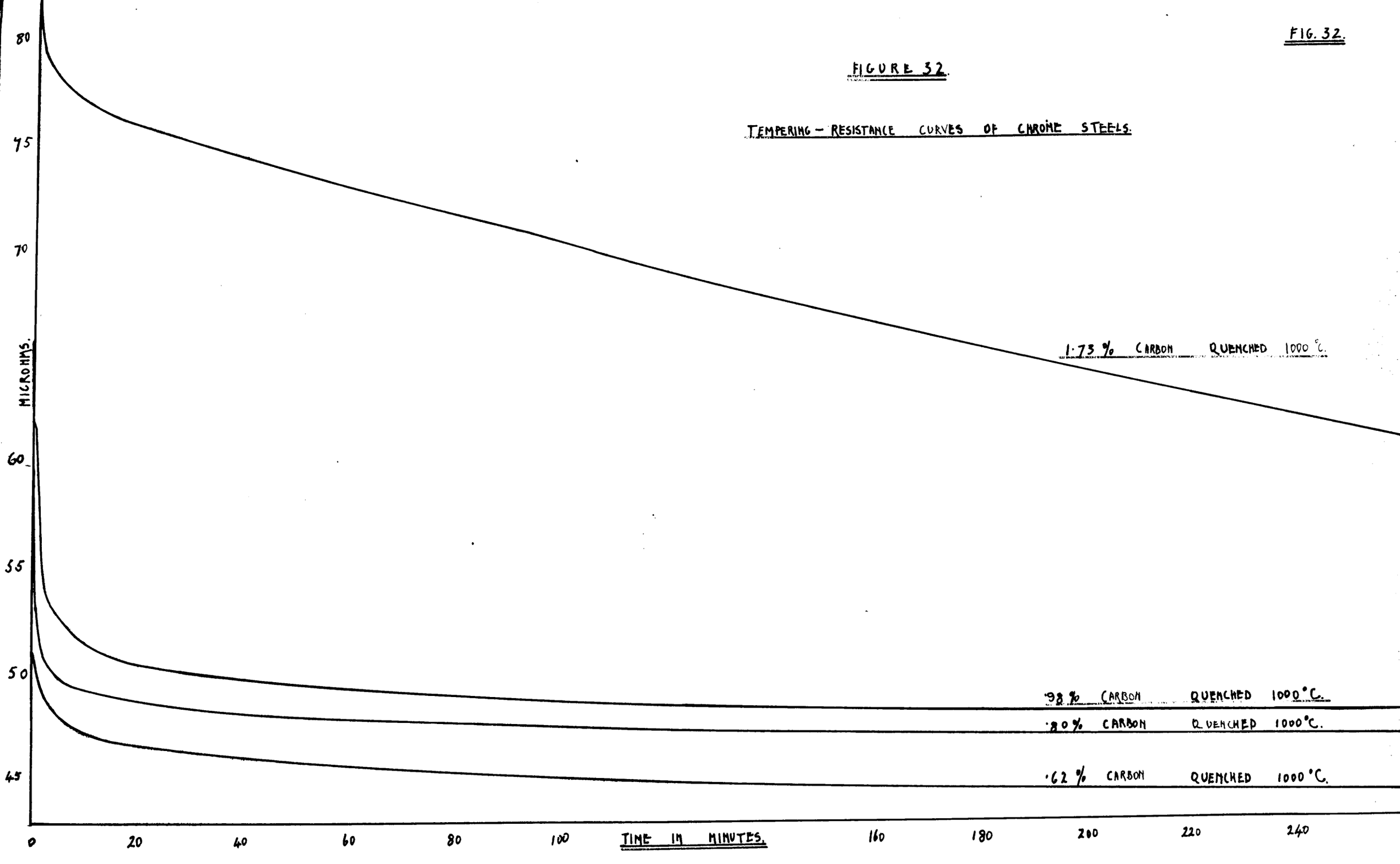


FIGURE 33.

TEMPERING-RESISTANCE CURVES OF CHROME STEELS.

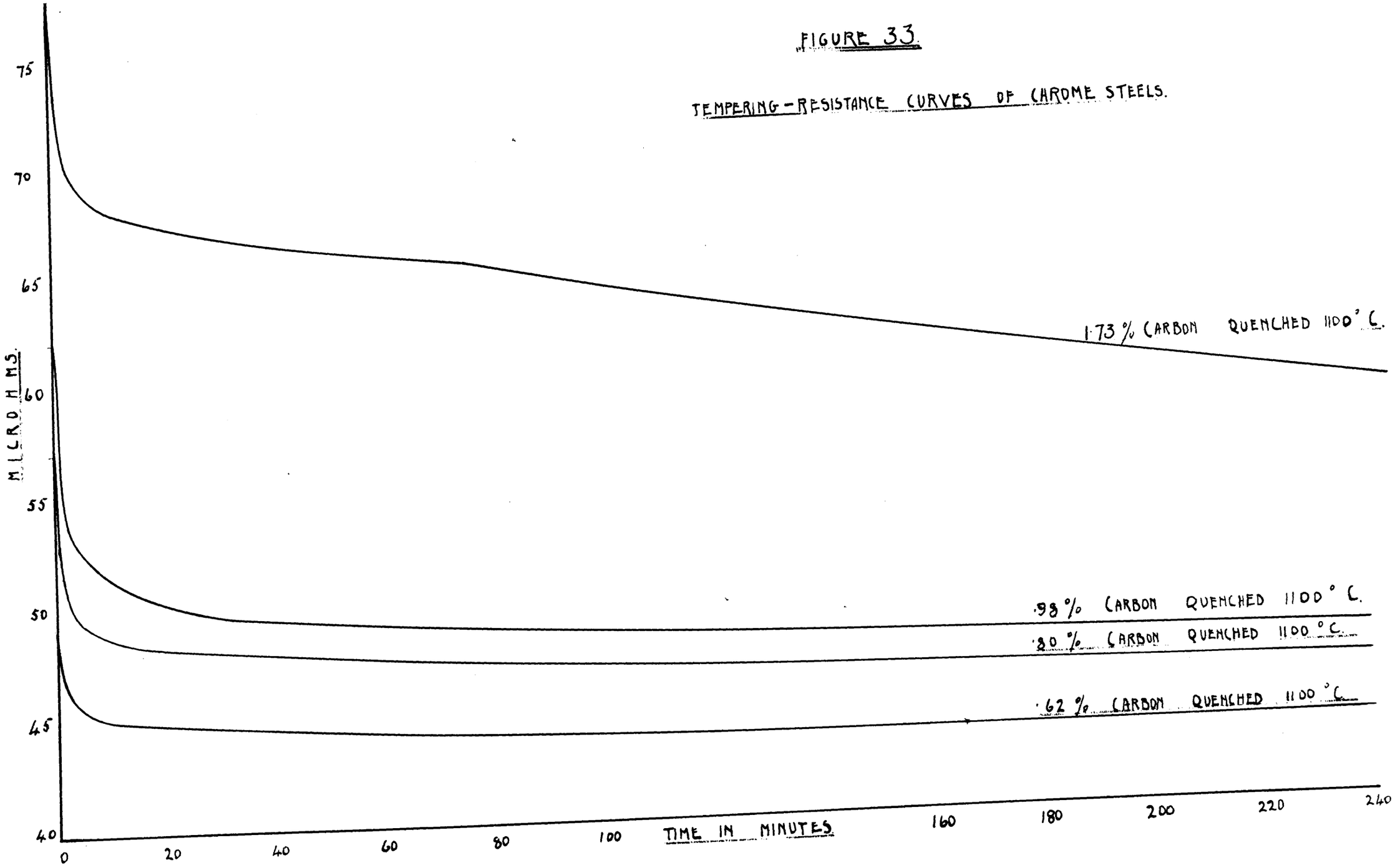


FIGURE 34

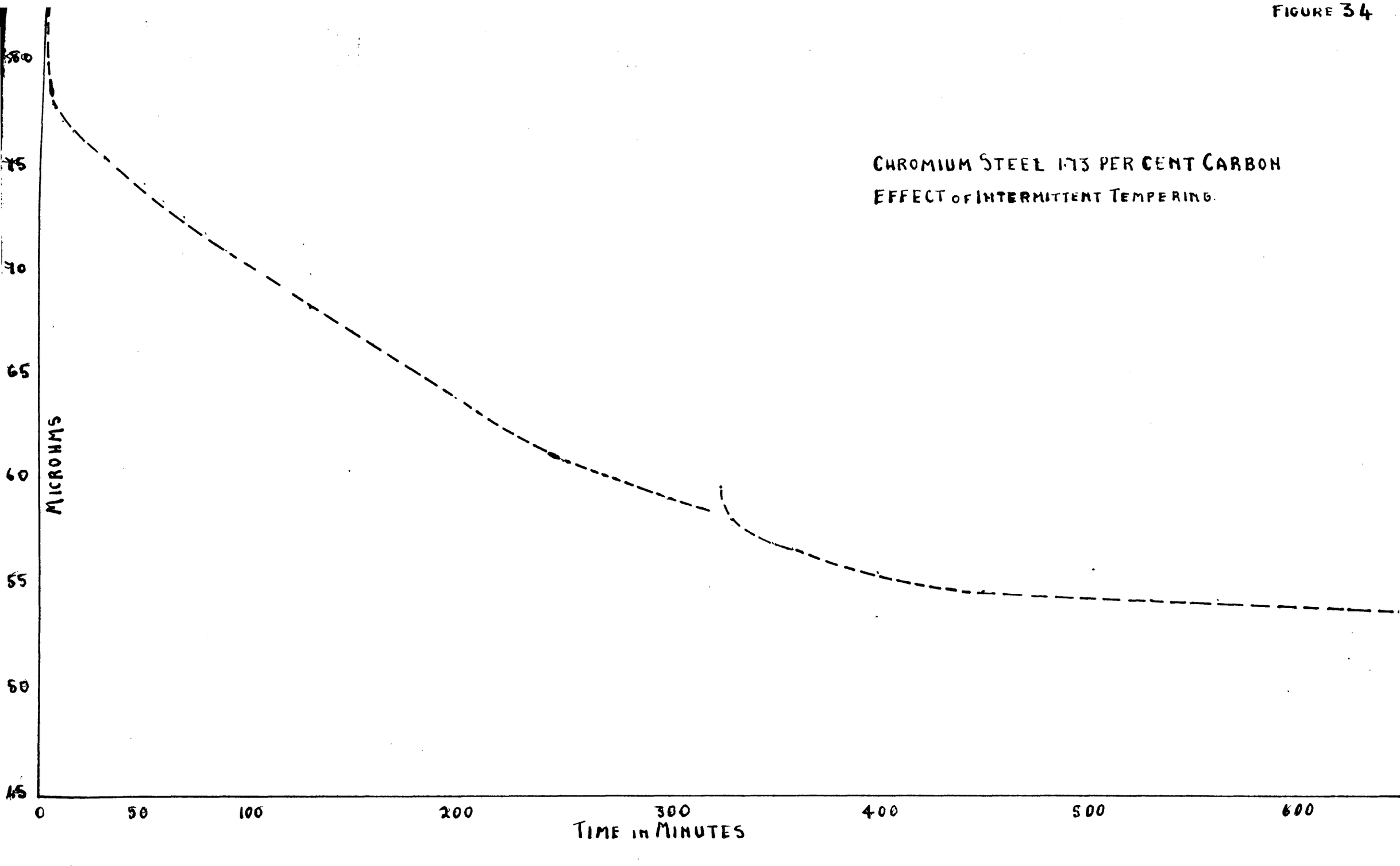

CHROMIUM STEEL 1·73 PER CENT CARBON
EFFECT OF INTERMITTENT TEMPERING.
MICROHMS
80
75
70
65
60
55
50
45
0
50
100
200
300
400
500
600
TIME IN MINUTES

FIGURE 35

TEMPERING - RESISTANCE CURVES OF NICKEL STEELS

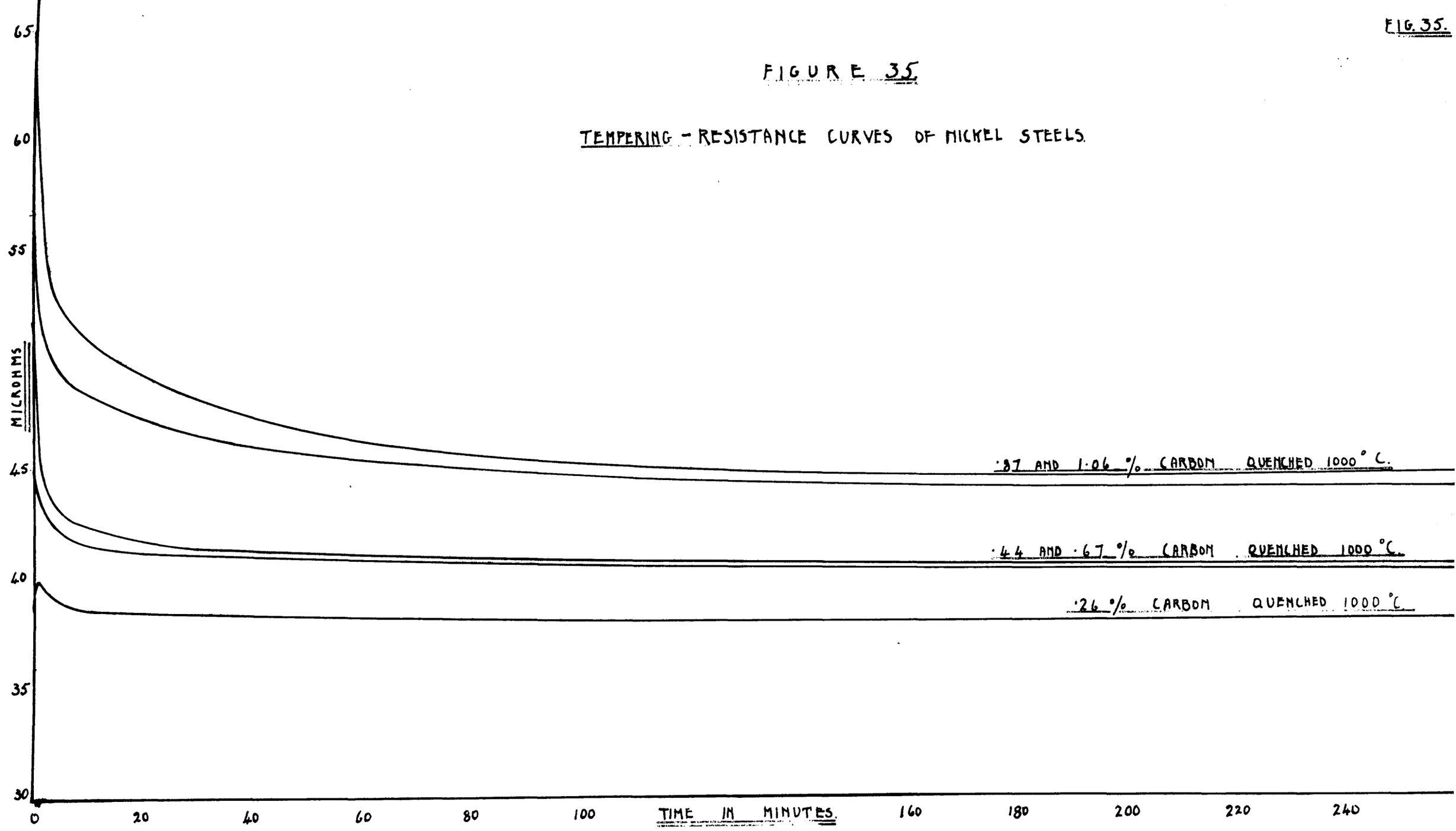
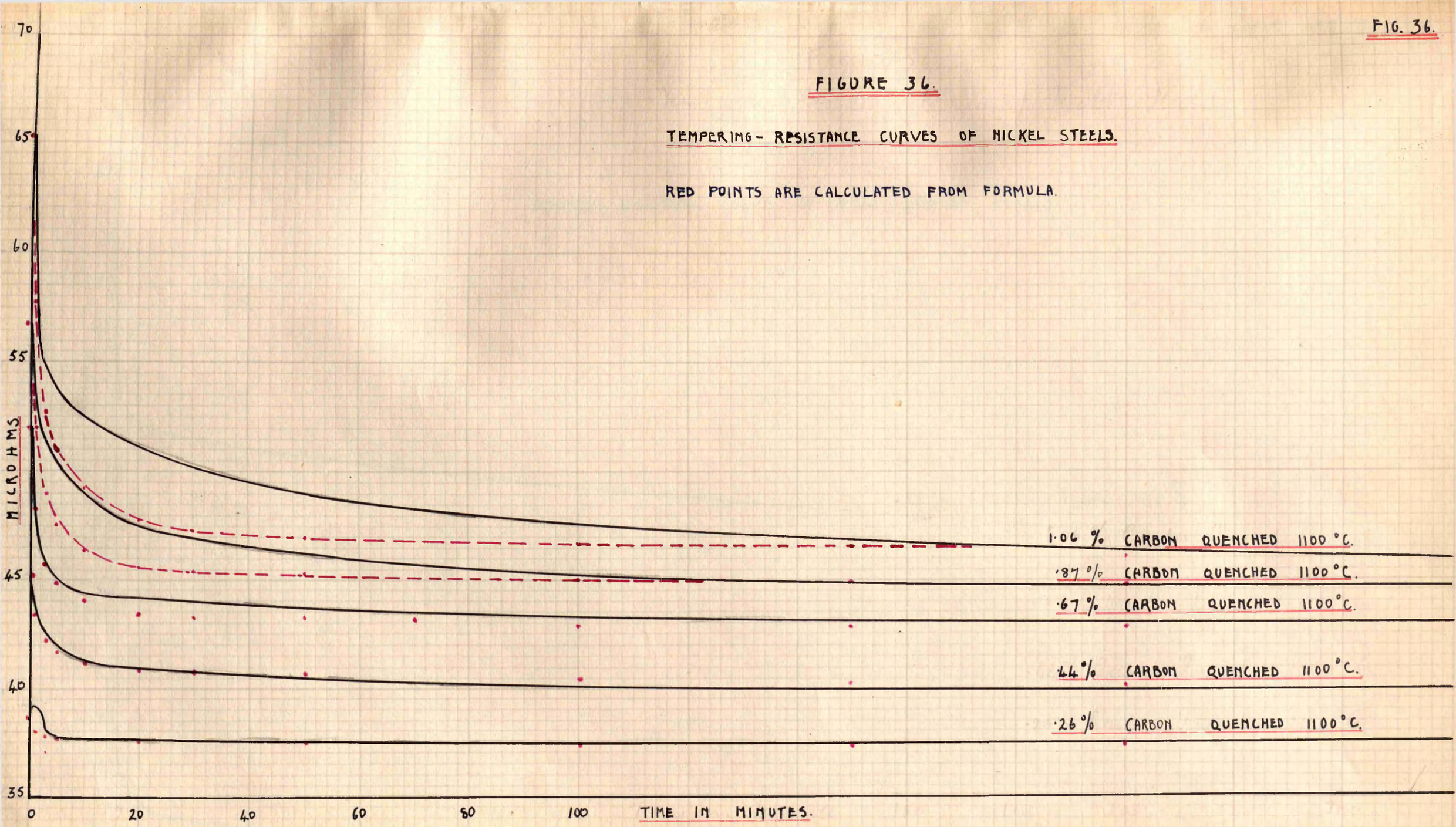


FIGURE 36.

TEMPERING - RESISTANCE CURVES OF NICKEL STEELS.

RED POINTS ARE CALCULATED FROM FORMULA.



1·50 % CARBON QUENCHED 1000°C.

FIGURE 37.

TEMPERING-RESISTANCE CURVES OF NICKEL-CHROMIUM STEELS

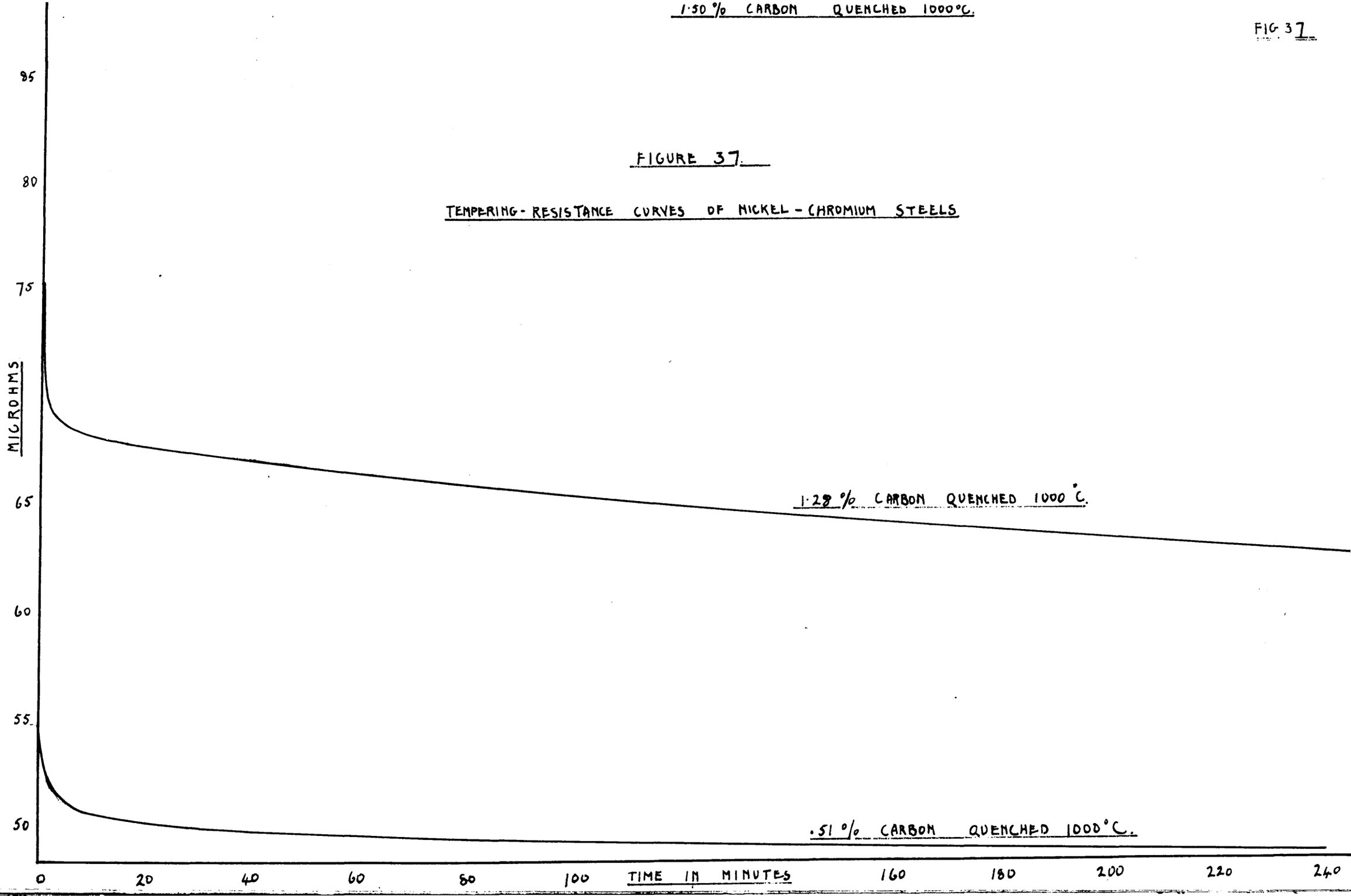


FIGURE 38

TEMPERING-RESISTANCE CURVES OF NICKEL-CHROMIUM STEELS.

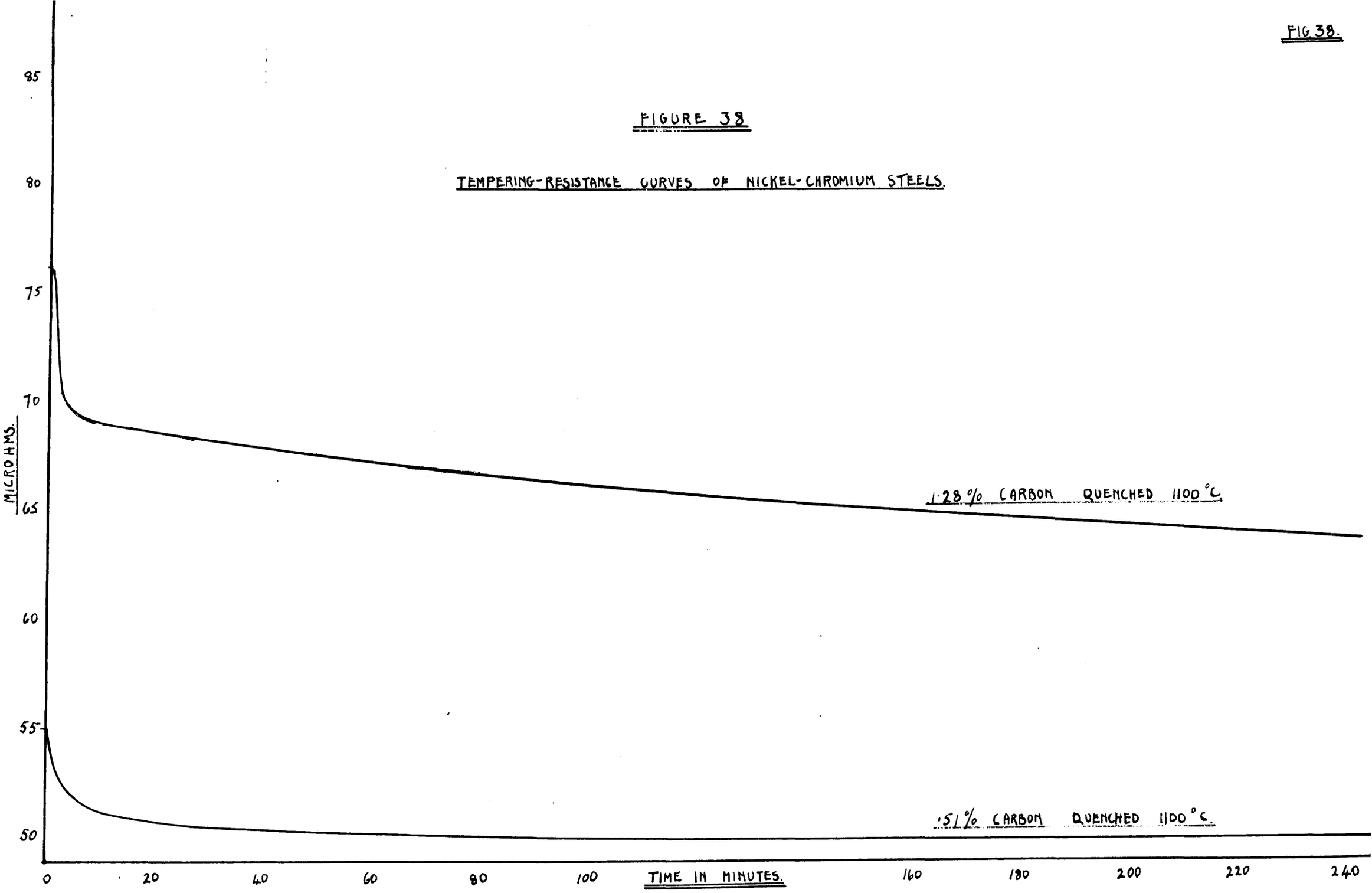


FIGURE 39

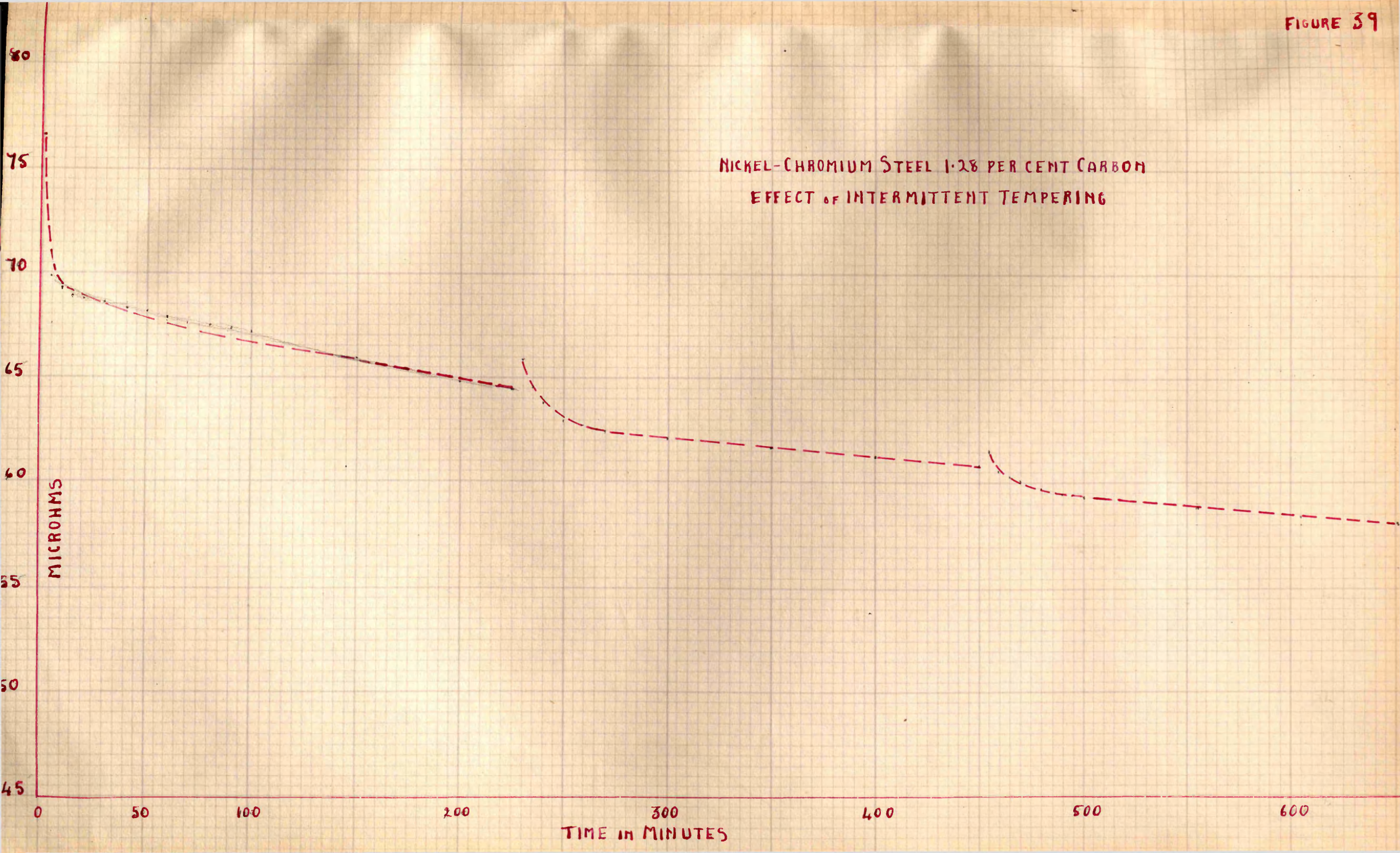

NICKEL-CHROMIUM STEEL 1·28 PER CENT CARBON
EFFECT OF INTERMITTENT TEMPERING
80
75
70
65
60
55
50
45
MICROHMS
0
50
100
200
300
400
500
600
TIME IN MINUTES

Figure 40.

Line AO Resistance at 240°C. of quenched carbon steels.

Line CO Resistance after 26 minutes tempering at 240°C.

From AO 1% carbon raises resistance 30 microhms.

From CO 1% carbon raises resistance 11.4 microhms.

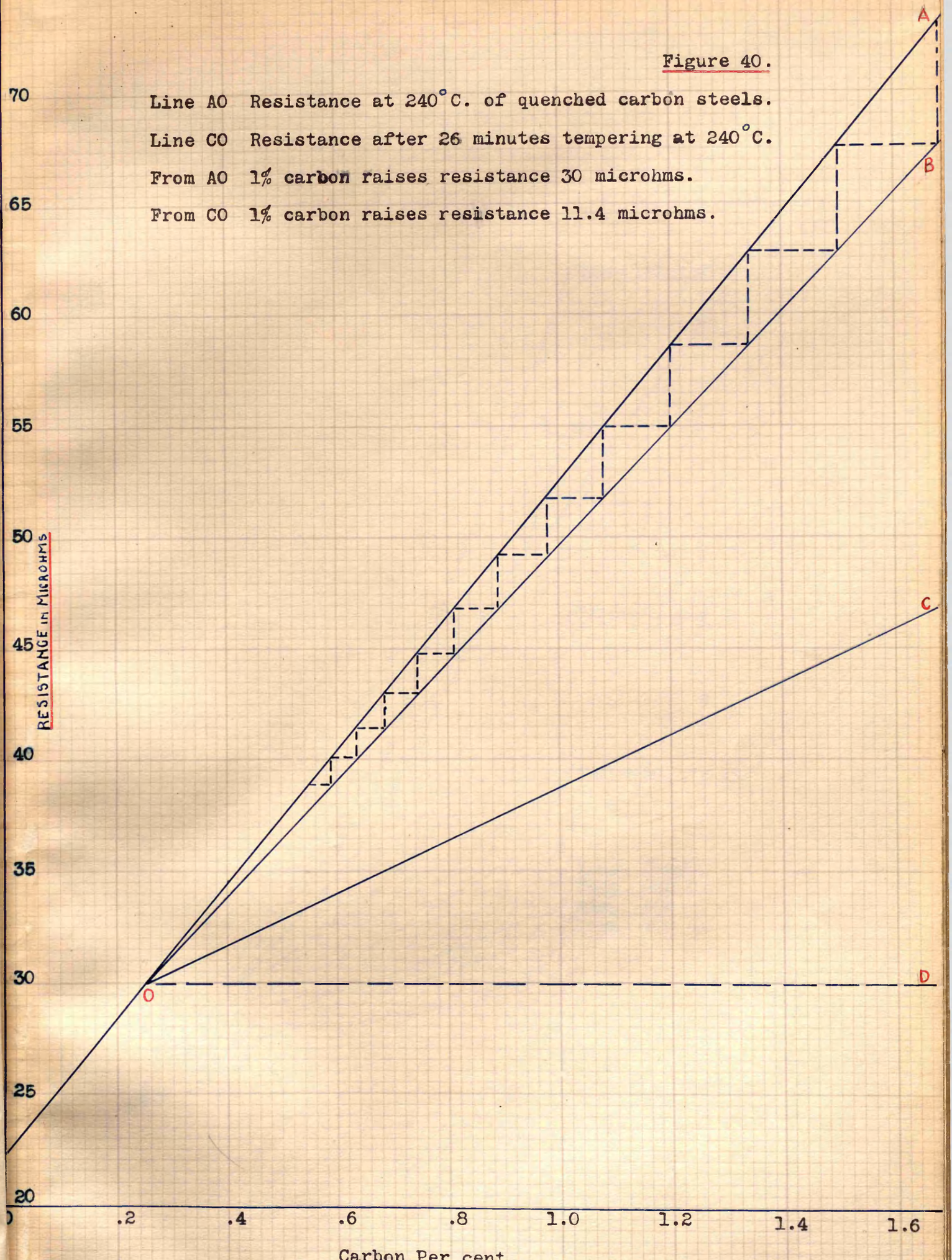


Figure 41.

Campbell's results showing effect of tempering at different temperatures on resistance of steels

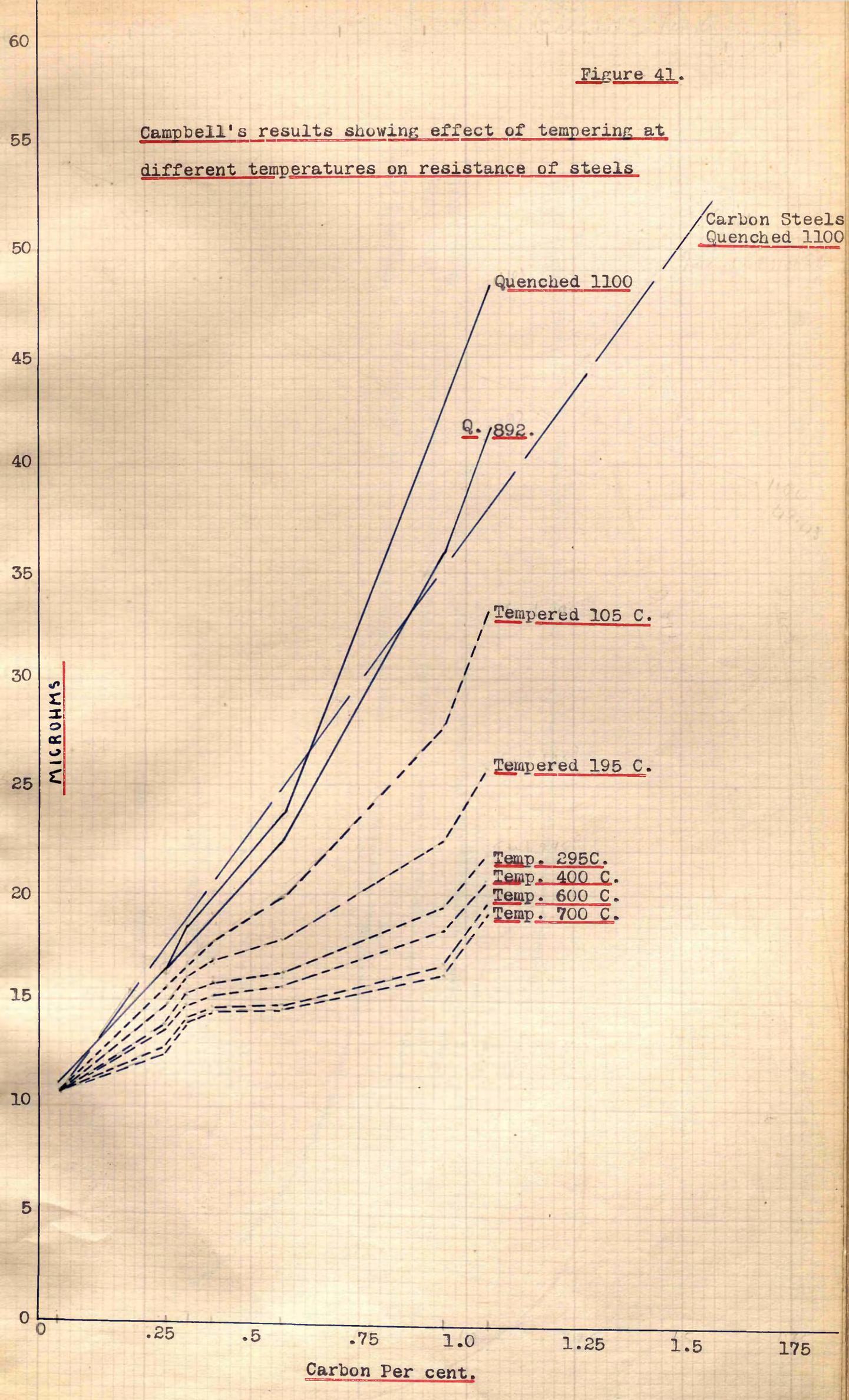


FIG. 42.

FIGURE 42

RATE AT WHICH RESISTANCE FALLS DUE TO SEPARATION OF CARBON FROM SOLUTION.

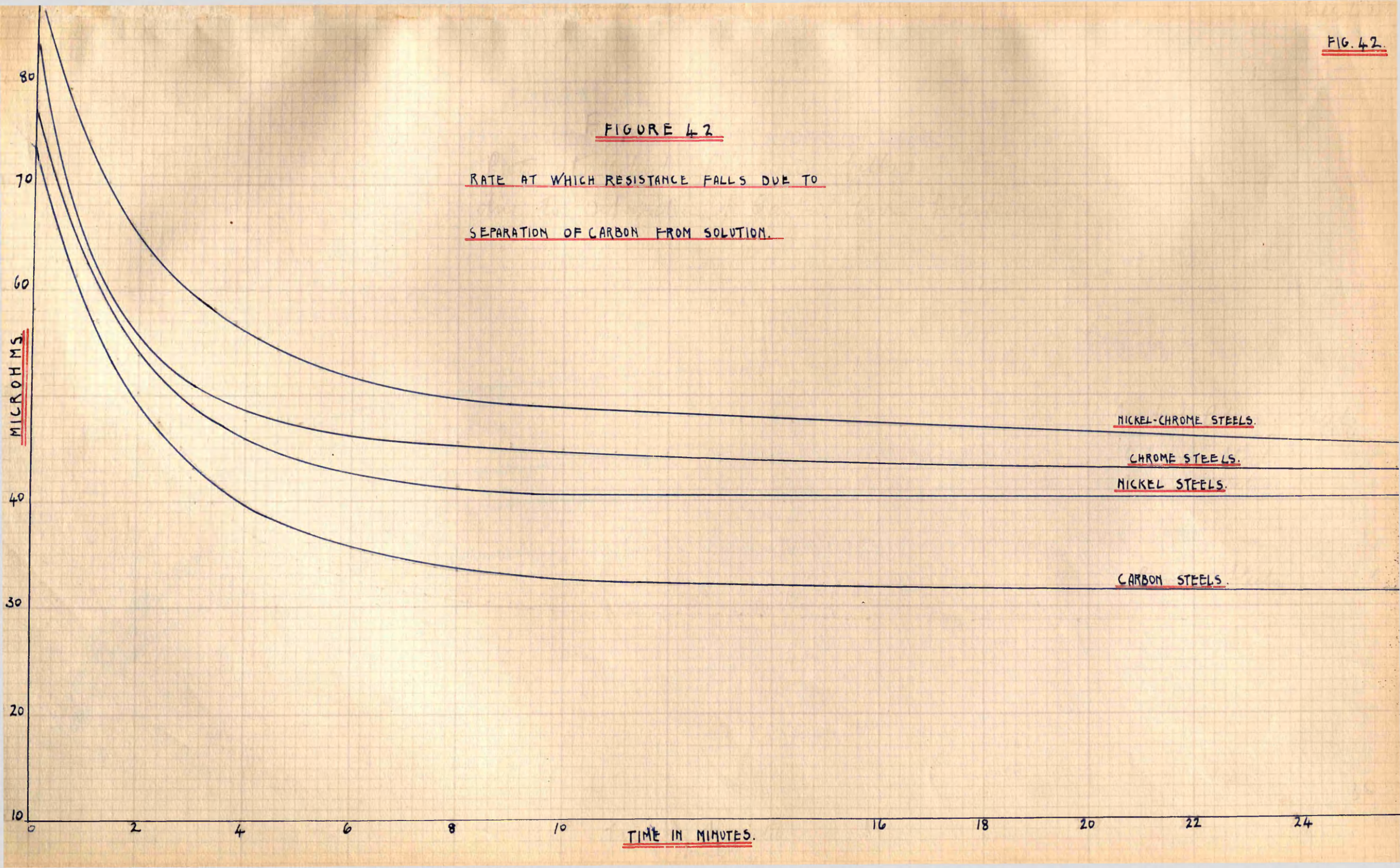
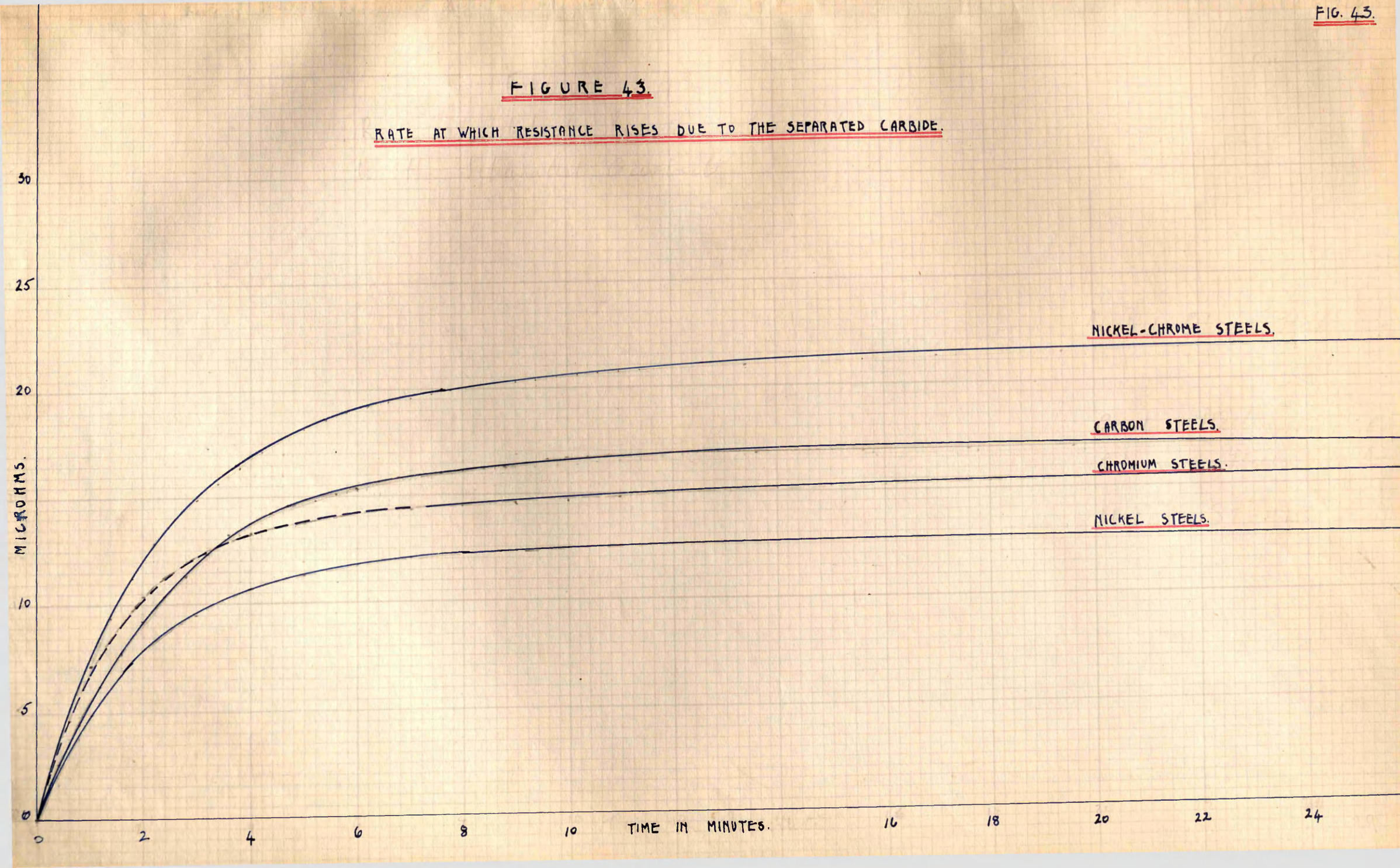


FIGURE 43.

RATE AT WHICH RESISTANCE RISES DUE TO THE SEPARATED CARBIDE.



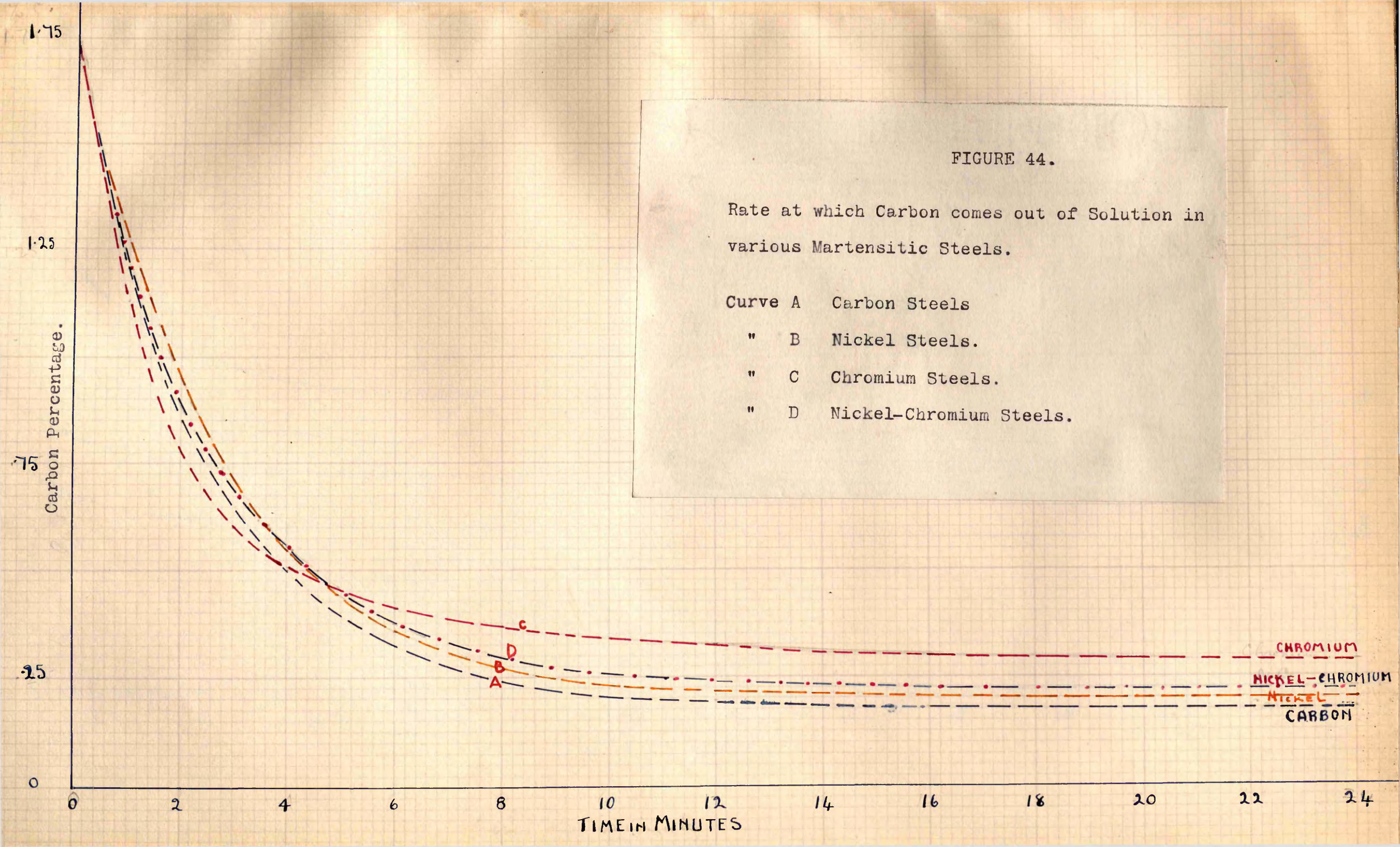


FIGURE 44.

Rate at which Carbon comes out of Solution in various Martensitic Steels.

Curve A Carbon Steels

" B Nickel Steels.

" C Chromium Steels.

" D Nickel-Chromium Steels.

FIG. 45

Figure 45.

Comparison between the Tempering Resistance Curves actually obtained for Carbon Steels and those calculated by means of Figures 42, 43 and 44.

Full lines are calculated curves.

Points are from actual Martensitic curves.

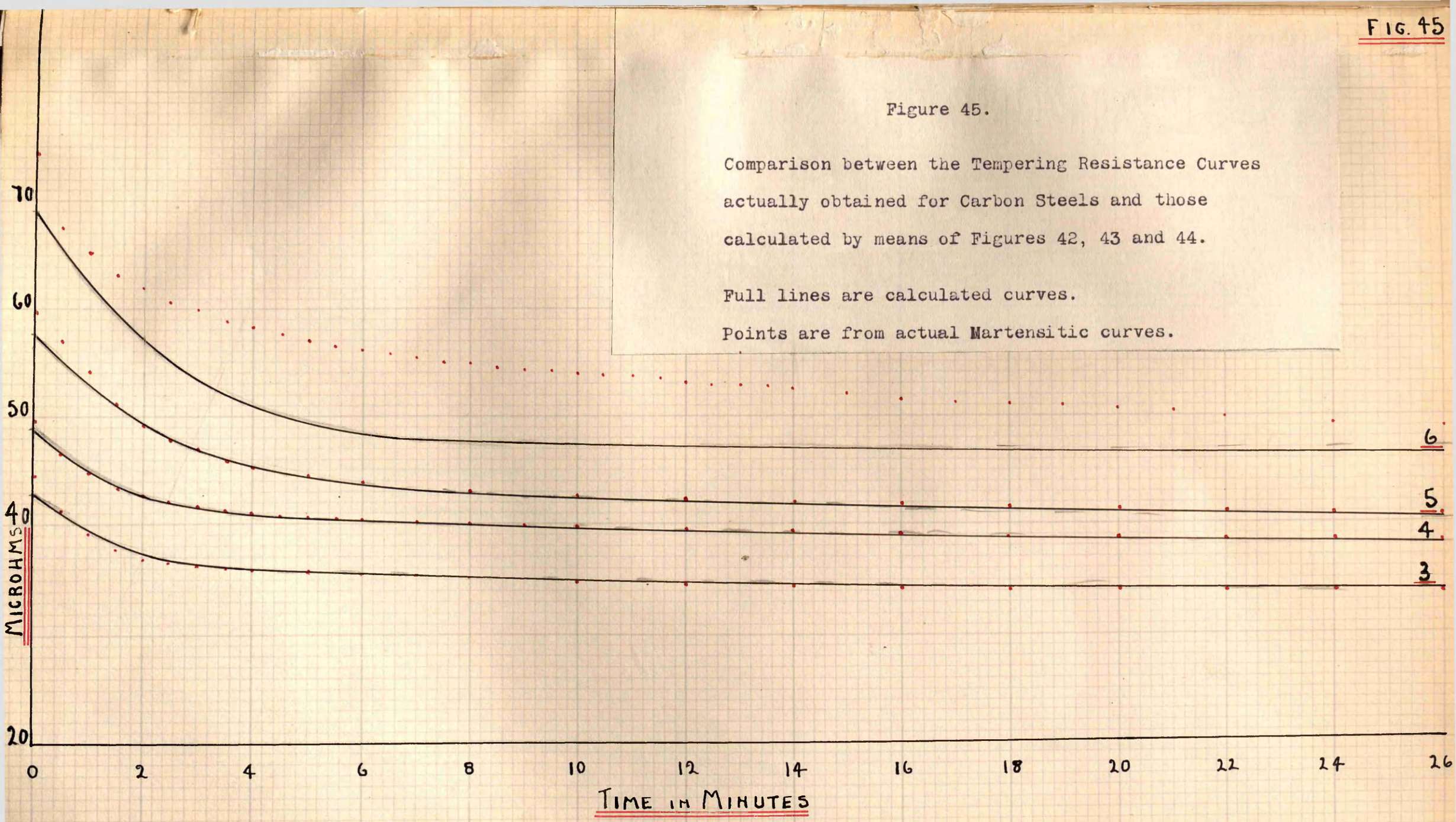
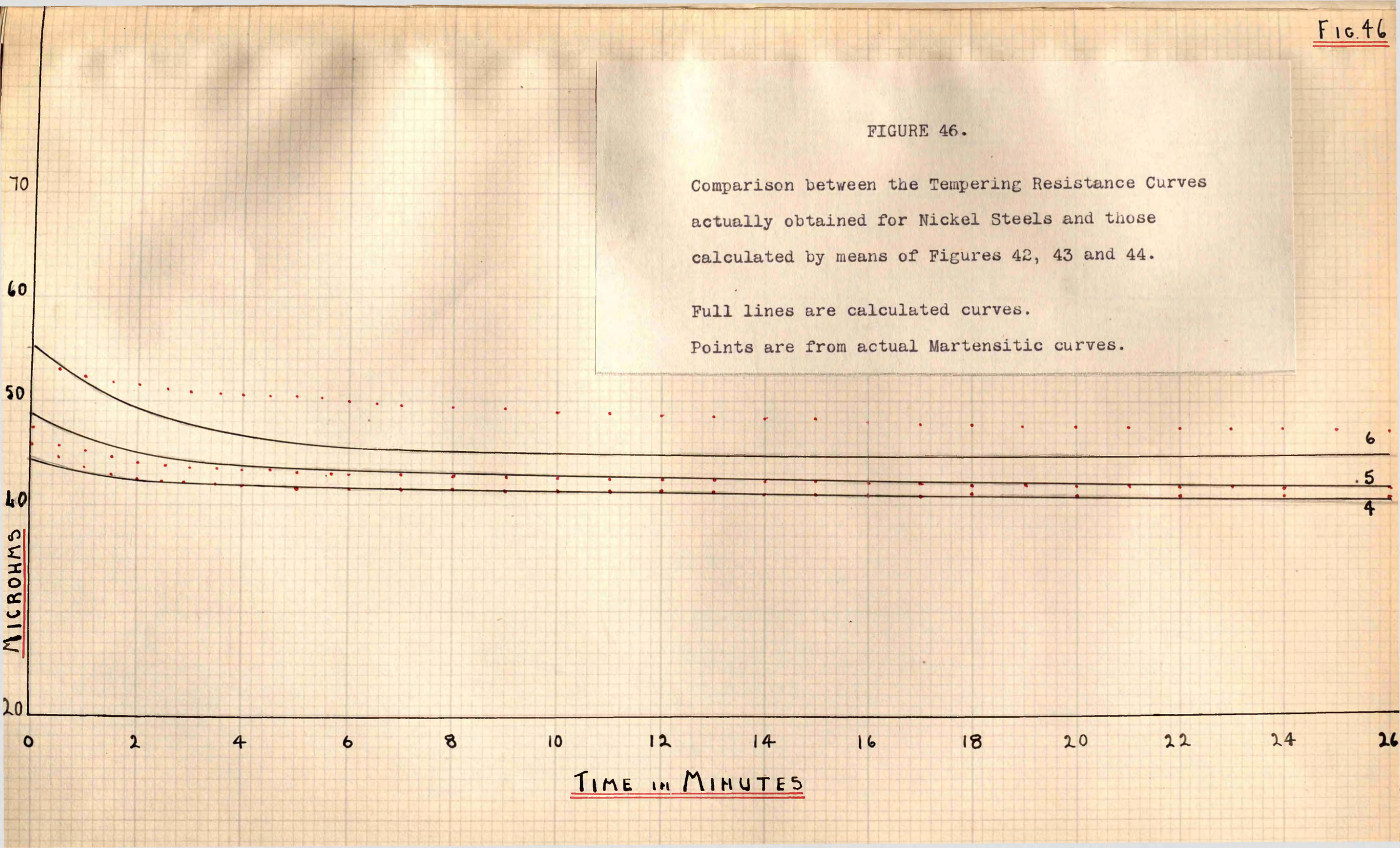


FIGURE 46.

Comparison between the Tempering Resistance Curves actually obtained for Nickel Steels and those calculated by means of Figures 42, 43 and 44.

Full lines are calculated curves.

Points are from actual Martensitic curves.



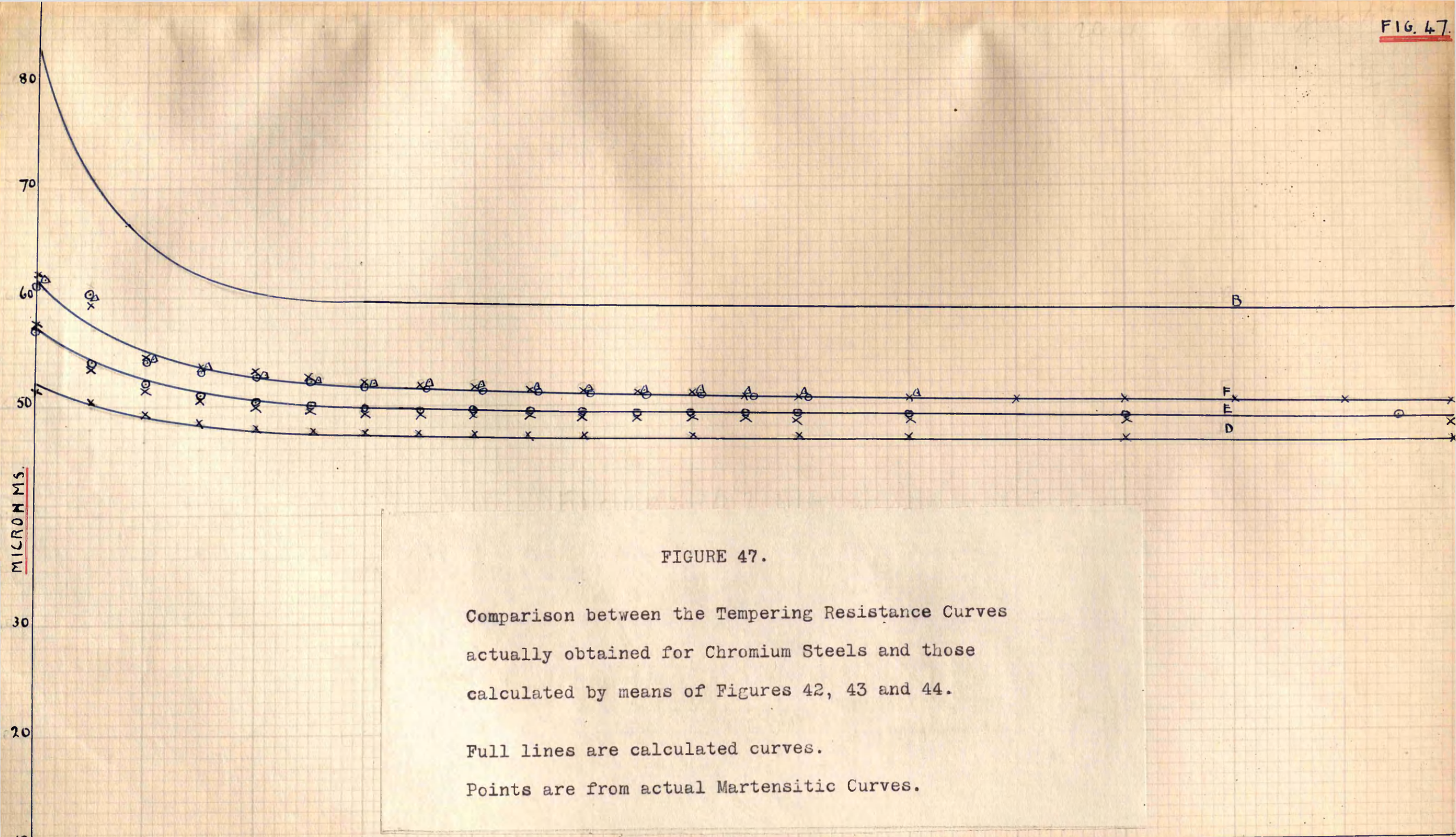


FIGURE 47.

Comparison between the Tempering Resistance Curves actually obtained for Chromium Steels and those calculated by means of Figures 42, 43 and 44.

Full lines are calculated curves.

Points are from actual Martensitic Curves.

FIG. 48

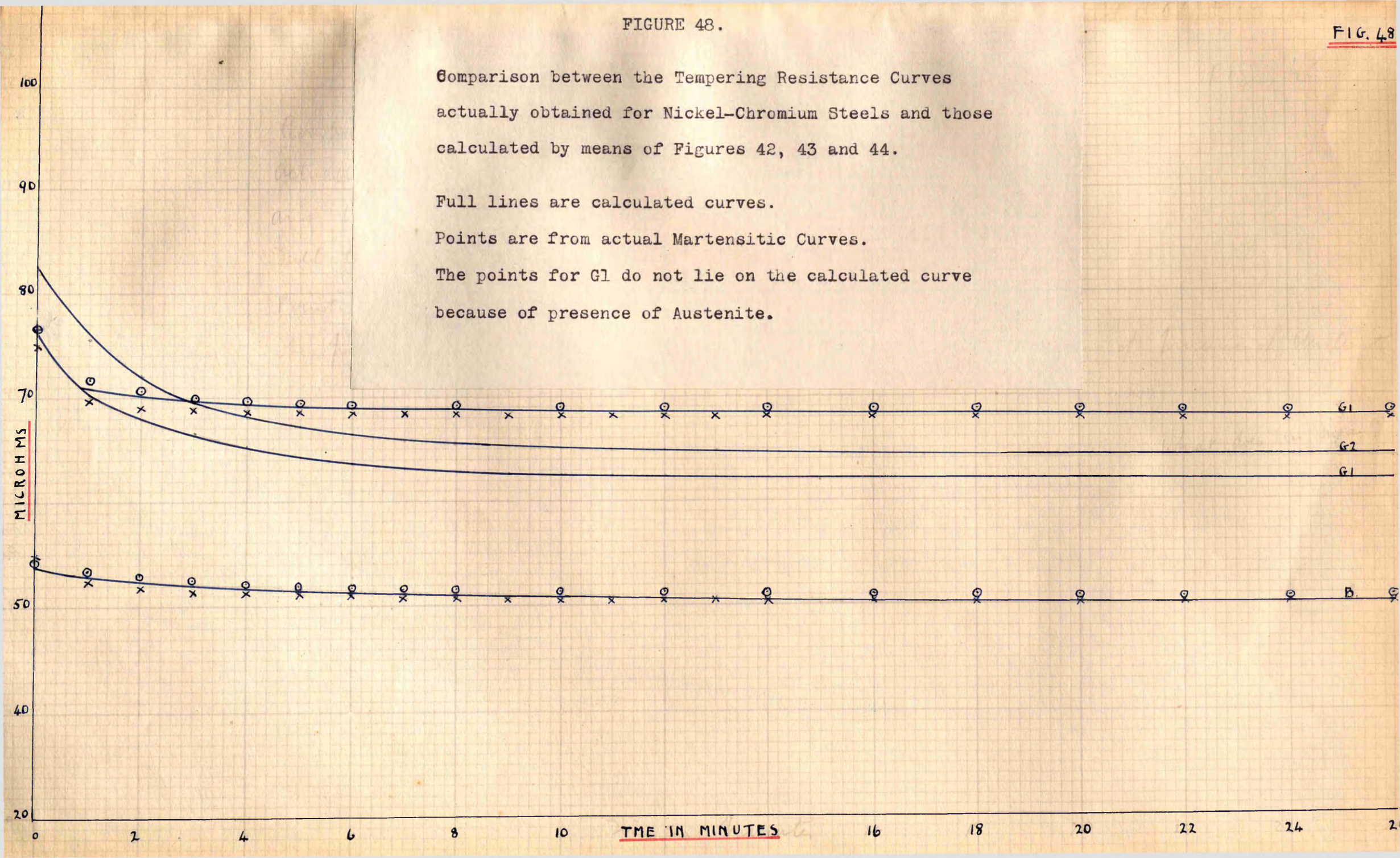
FIGURE 48.

Comparison between the Tempering Resistance Curves actually obtained for Nickel-Chromium Steels and those calculated by means of Figures 42, 43 and 44.

Full lines are calculated curves.

Points are from actual Martensitic Curves.

The points for G1 do not lie on the calculated curve because of presence of Austenite.



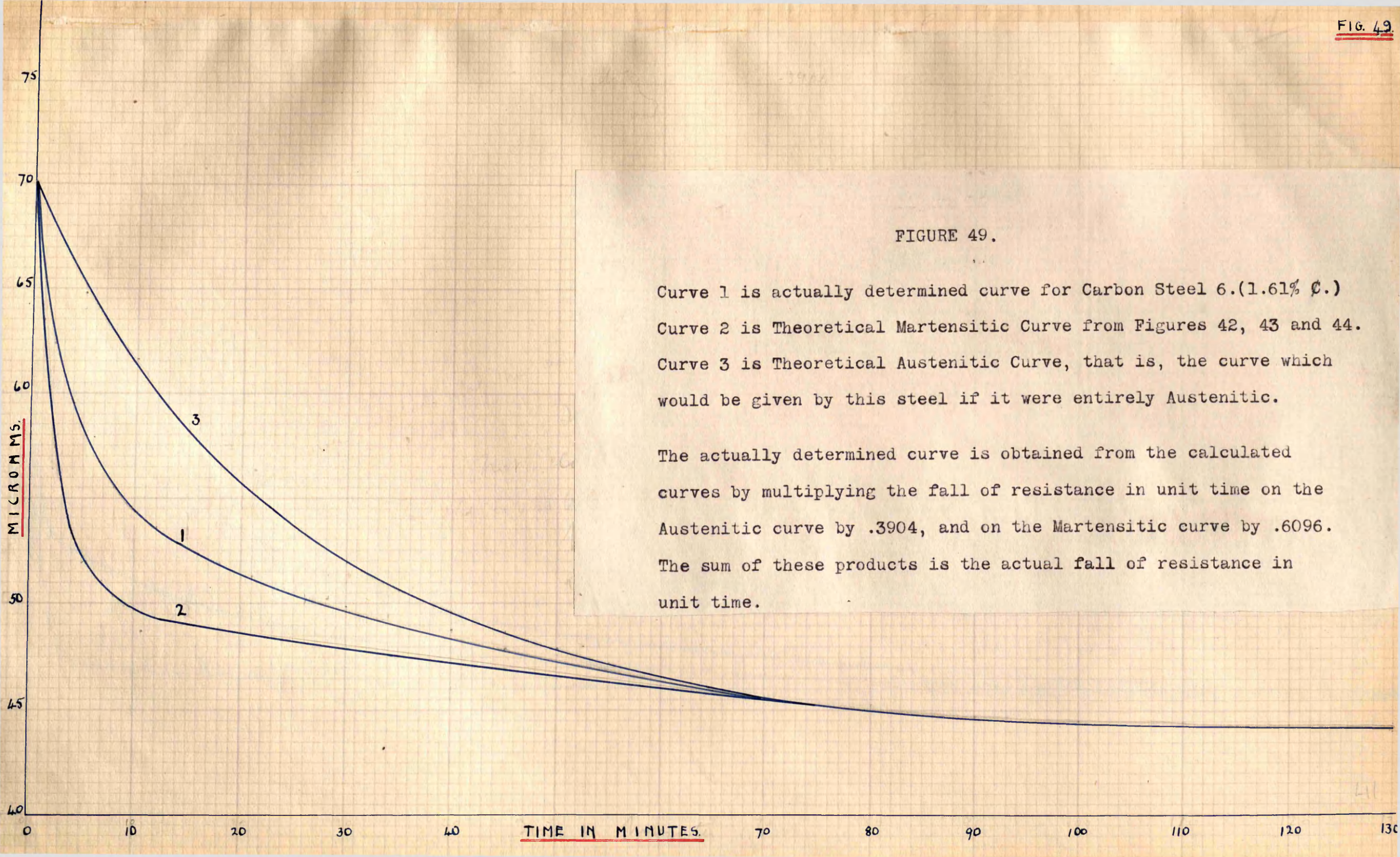


FIGURE 49.

Curve 1 is actually determined curve for Carbon Steel 6.(1.61% ₵.)
Curve 2 is Theoretical Martensitic Curve from Figures 42, 43 and 44.
Curve 3 is Theoretical Austenitic Curve, that is, the curve which would be given by this steel if it were entirely Austenitic.

The actually determined curve is obtained from the calculated curves by multiplying the fall of resistance in unit time on the Austenitic curve by .3904, and on the Martensitic curve by .6096. The sum of these products is the actual fall of resistance in unit time.

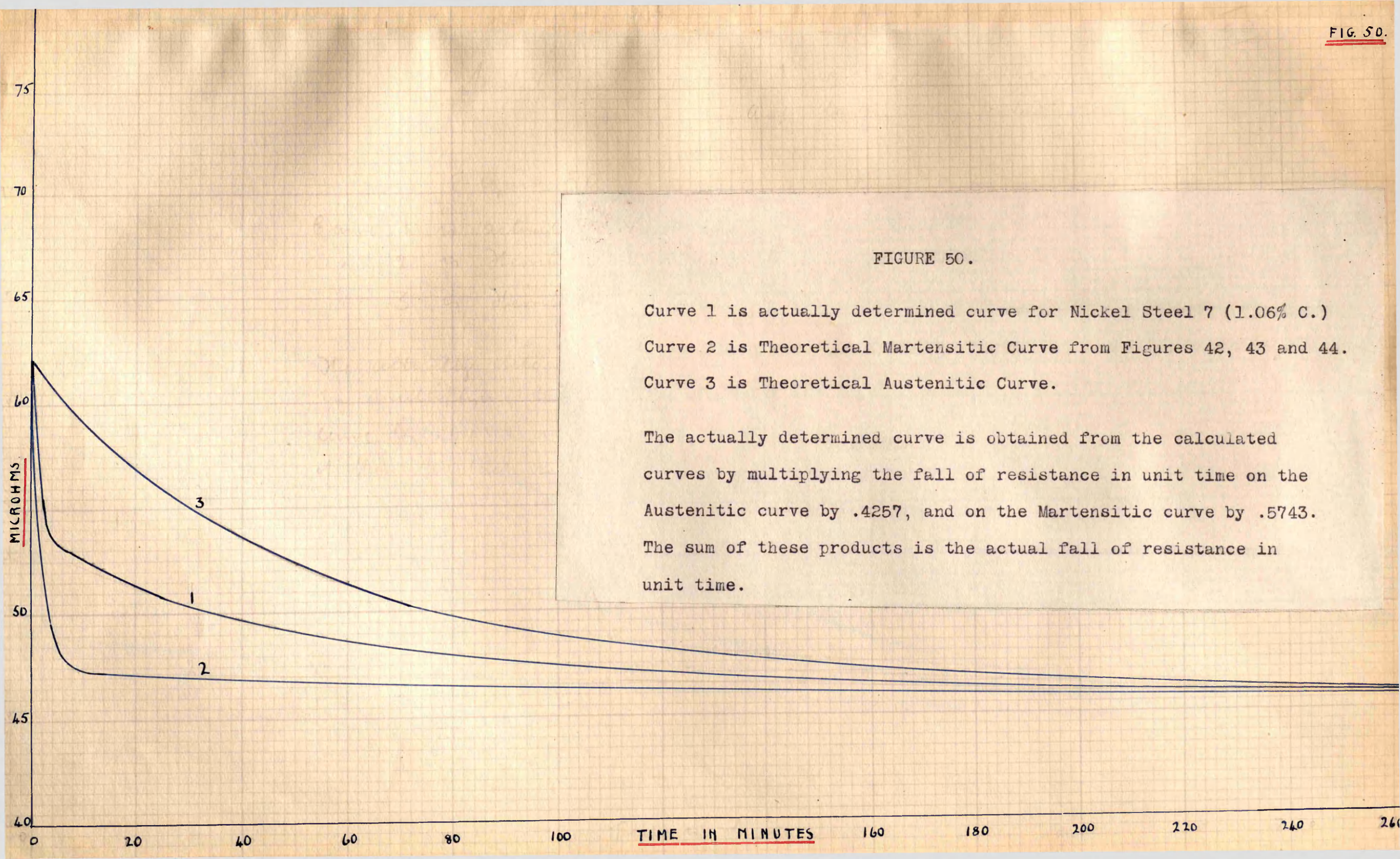


FIGURE 50.

Curve 1 is actually determined curve for Nickel Steel 7 (1.06% C.)

Curve 2 is Theoretical Martensitic Curve from Figures 42, 43 and 44.

Curve 3 is Theoretical Austenitic Curve.

The actually determined curve is obtained from the calculated curves by multiplying the fall of resistance in unit time on the Austenitic curve by .4257, and on the Martensitic curve by .5743. The sum of these products is the actual fall of resistance in unit time.

FIGURE 51.

Curve 1 is actually determined curve for Chromium Steel B (1.73%C.)
Quenched 1100 C.

Curve 2 is Theoretical Martensitic Curve from Figures 42, 43 and 44.

Curve 3 is Theoretical Austenitic Curve.

Fall in unit time on Austenitic Curve multiplied by .7859 plus fall in unit time of Martensitic curve multiplied by .2140, equals fall in unit time on actual curve.

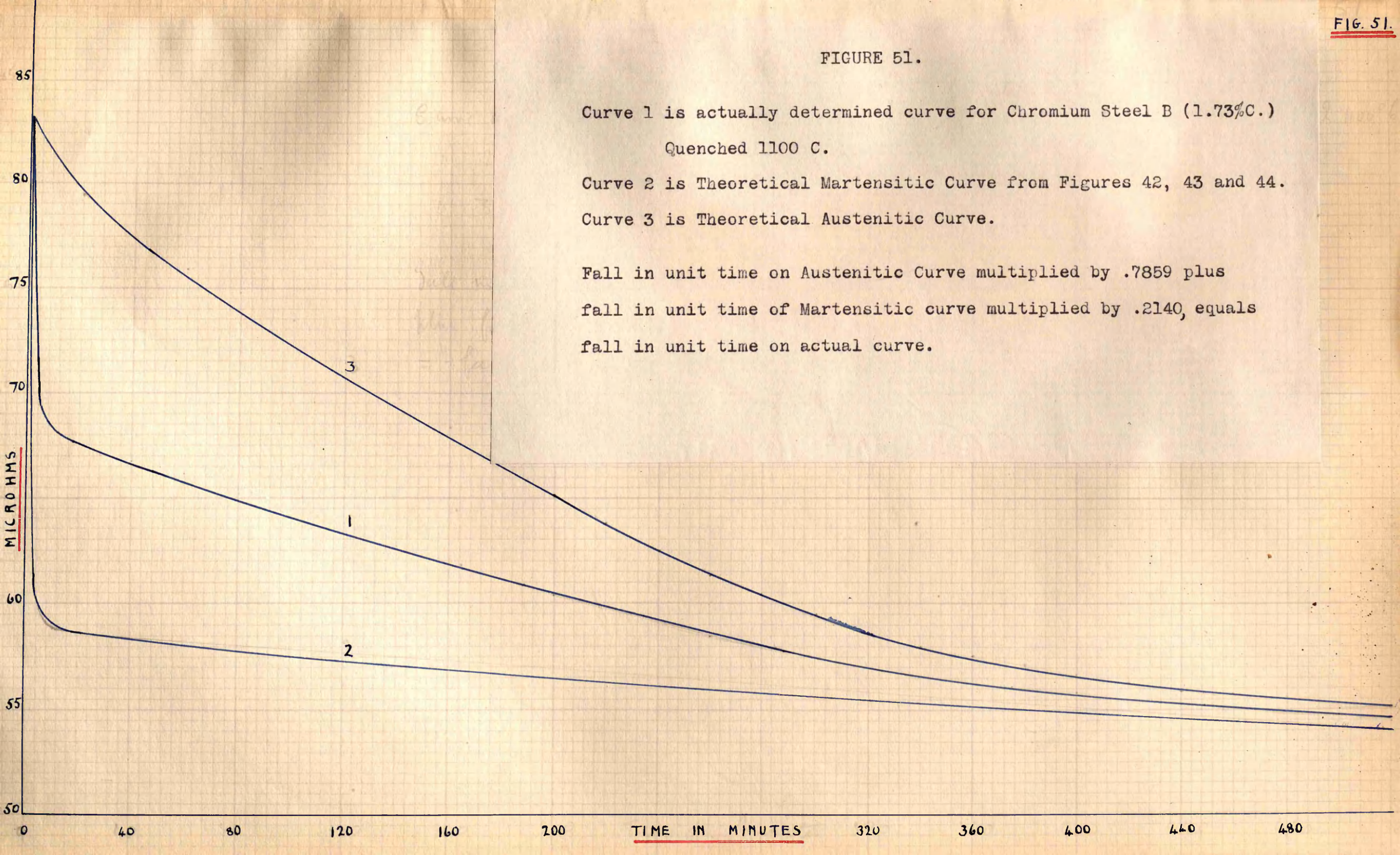


FIG. 52.

FIGURE 52.

Curve 1 is actually determined curve for Chromium Steel B. Quen.1000 C.

Curve 2 is Theoretical Martensitic Curve from Figures 42, 43 and 44.

Curve 3 is Theoretical Austenitic Curve.

Fall in unit time on curve 3 multiplied by .7859 plus fall in unit time on curve 2 multiplied by .2140, gives fall in unit time on curve 1.

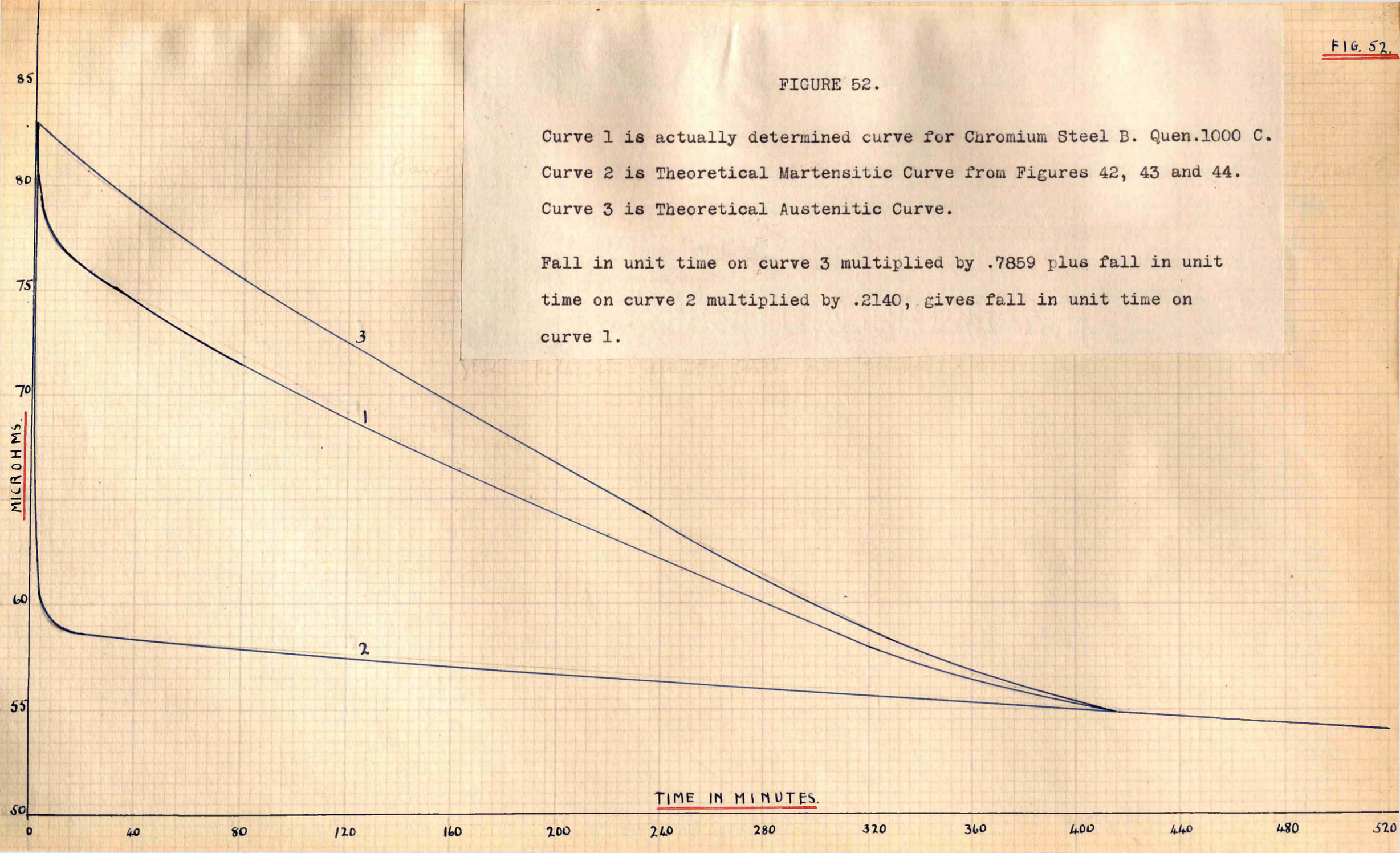


FIGURE 53.

Curve 1 is actuall determined curve for Nickel-Chrome Steel Gl.(1.28%C.)

Curve 2 is Theoretical Martensitic Curve from Figures 42, 43 and 44.

Curve 3 is Theoretical Austenitic Curve.

Fall of resistance in unit time on Curve 3 multiplied by .5572 plus fall in unit time on Curve 2 multiplied by .4429, equals fall in unit time on Curve 1.

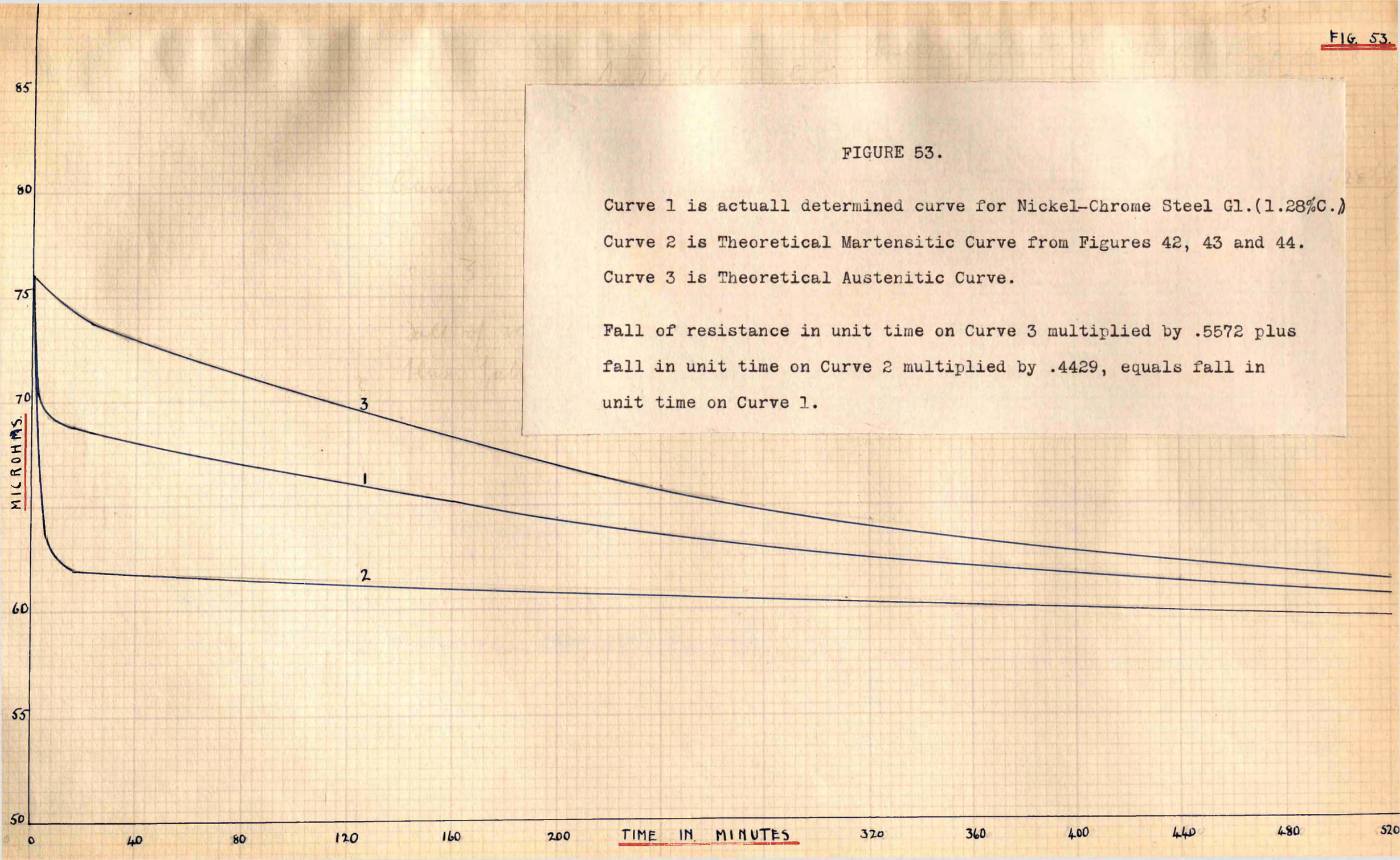
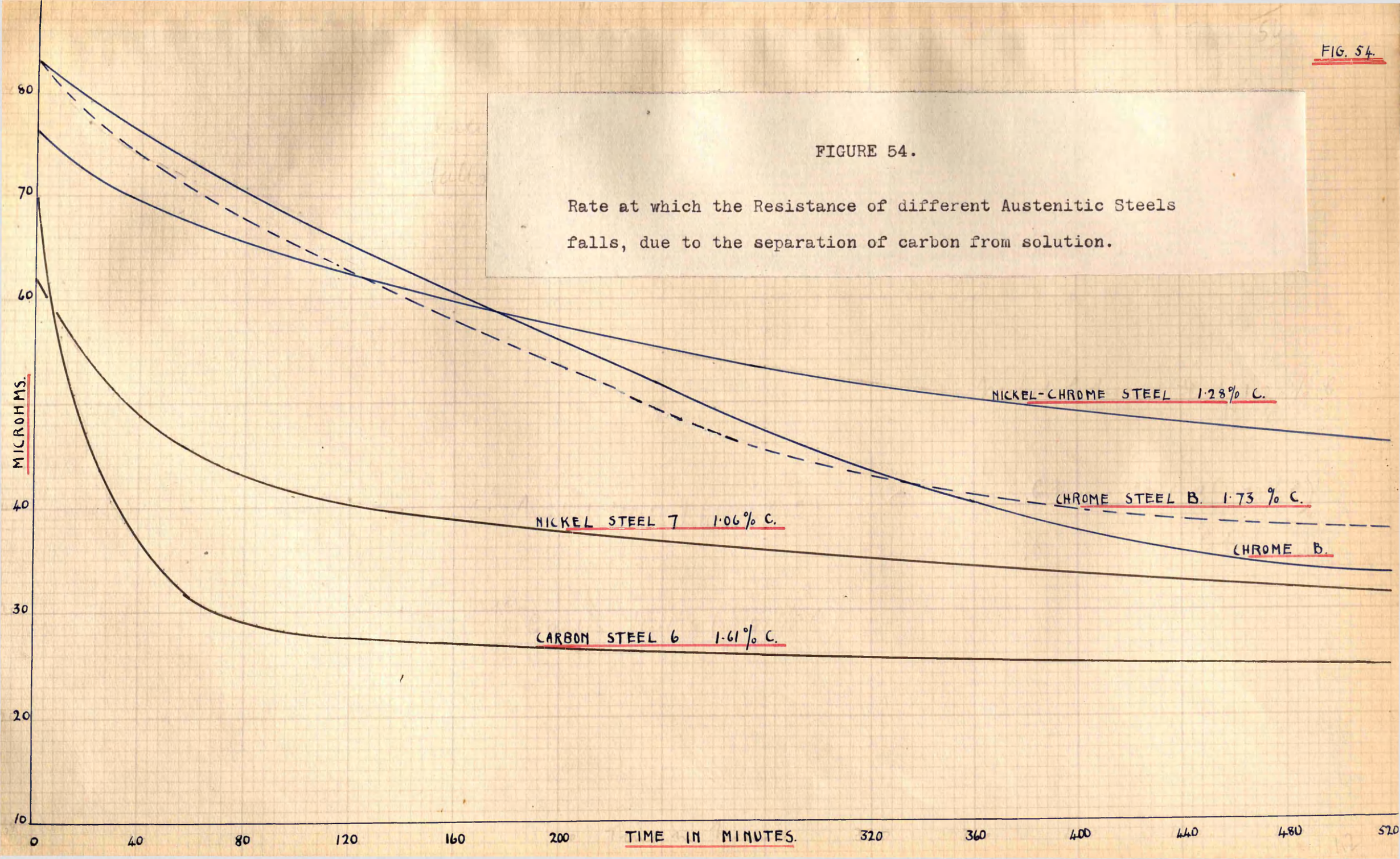


FIG. 54.

FIGURE 54.

Rate at which the Resistance of different Austenitic Steels falls, due to the separation of carbon from solution.



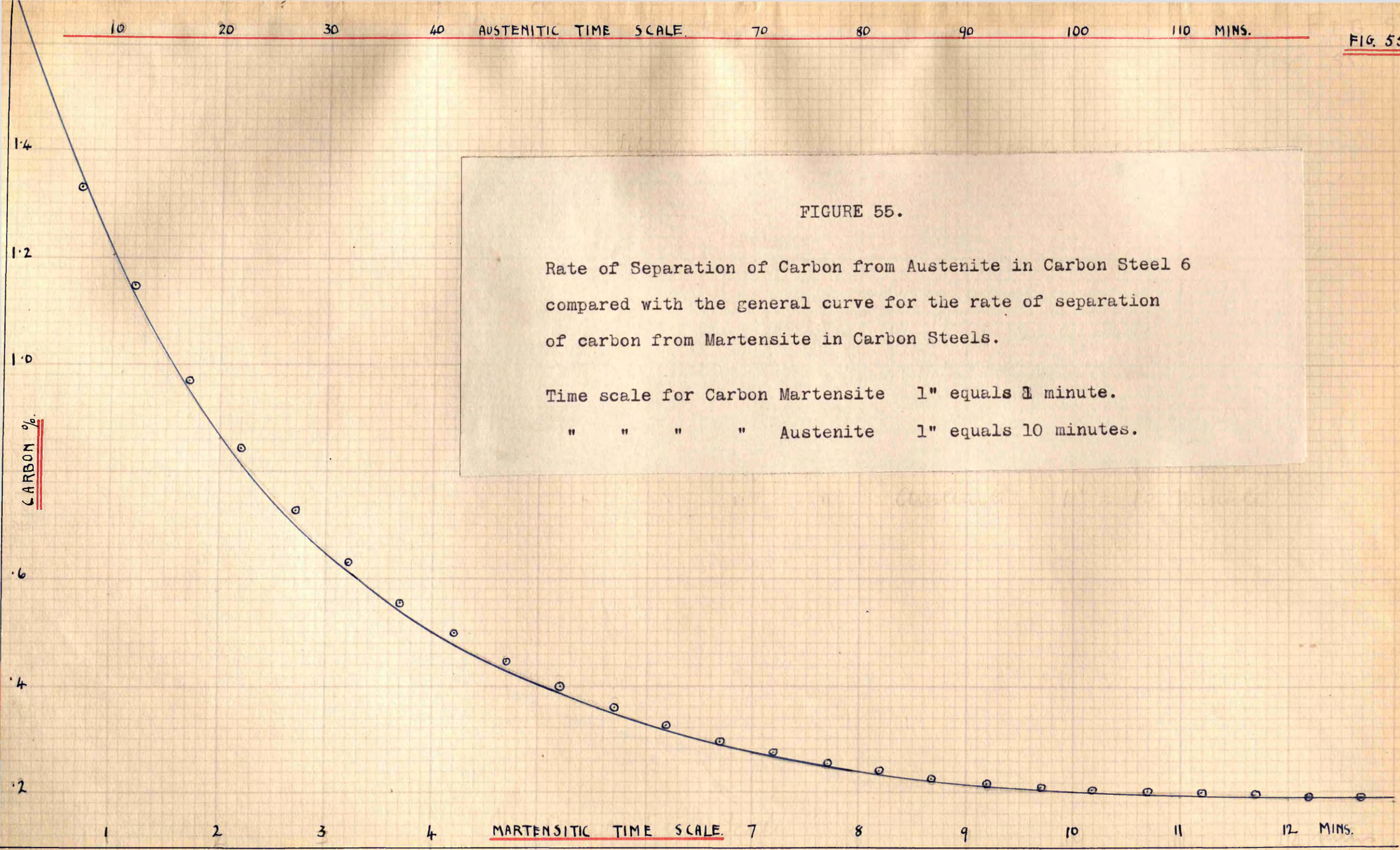


FIGURE 55.

Rate of Separation of Carbon from Austenite in Carbon Steel 6 compared with the general curve for the rate of separation of carbon from Martensite in Carbon Steels.

Time scale for Carbon Martensite 1" equals 1 minute.

" " " " Austenite 1" equals 10 minutes.

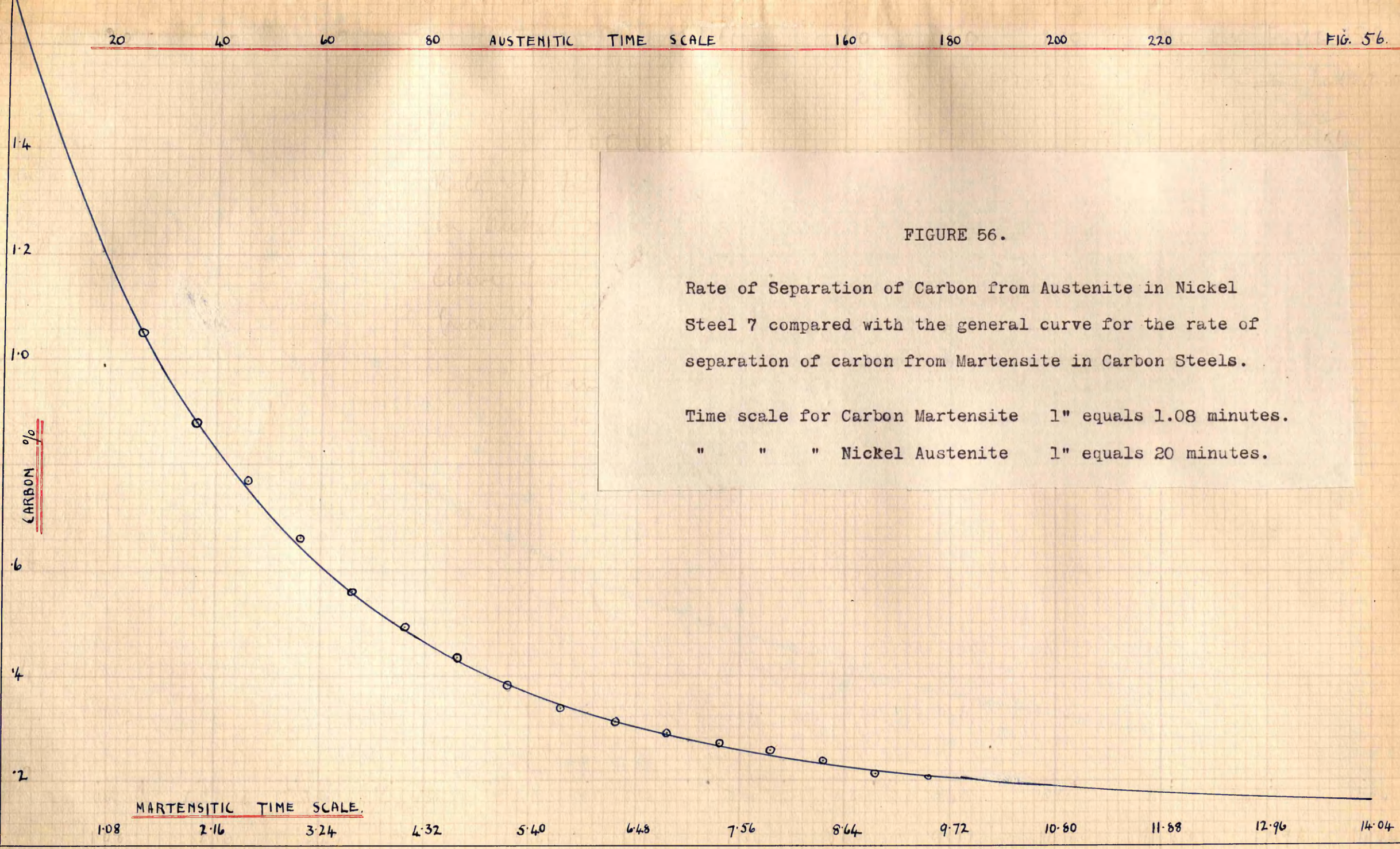


FIGURE 56.

Rate of Separation of Carbon from Austenite in Nickel Steel 7 compared with the general curve for the rate of separation of carbon from Martensite in Carbon Steels.

Time scale for Carbon Martensite 1" equals 1.08 minutes.

" " " Nickel Austenite 1" equals 20 minutes.

FIG. 57.

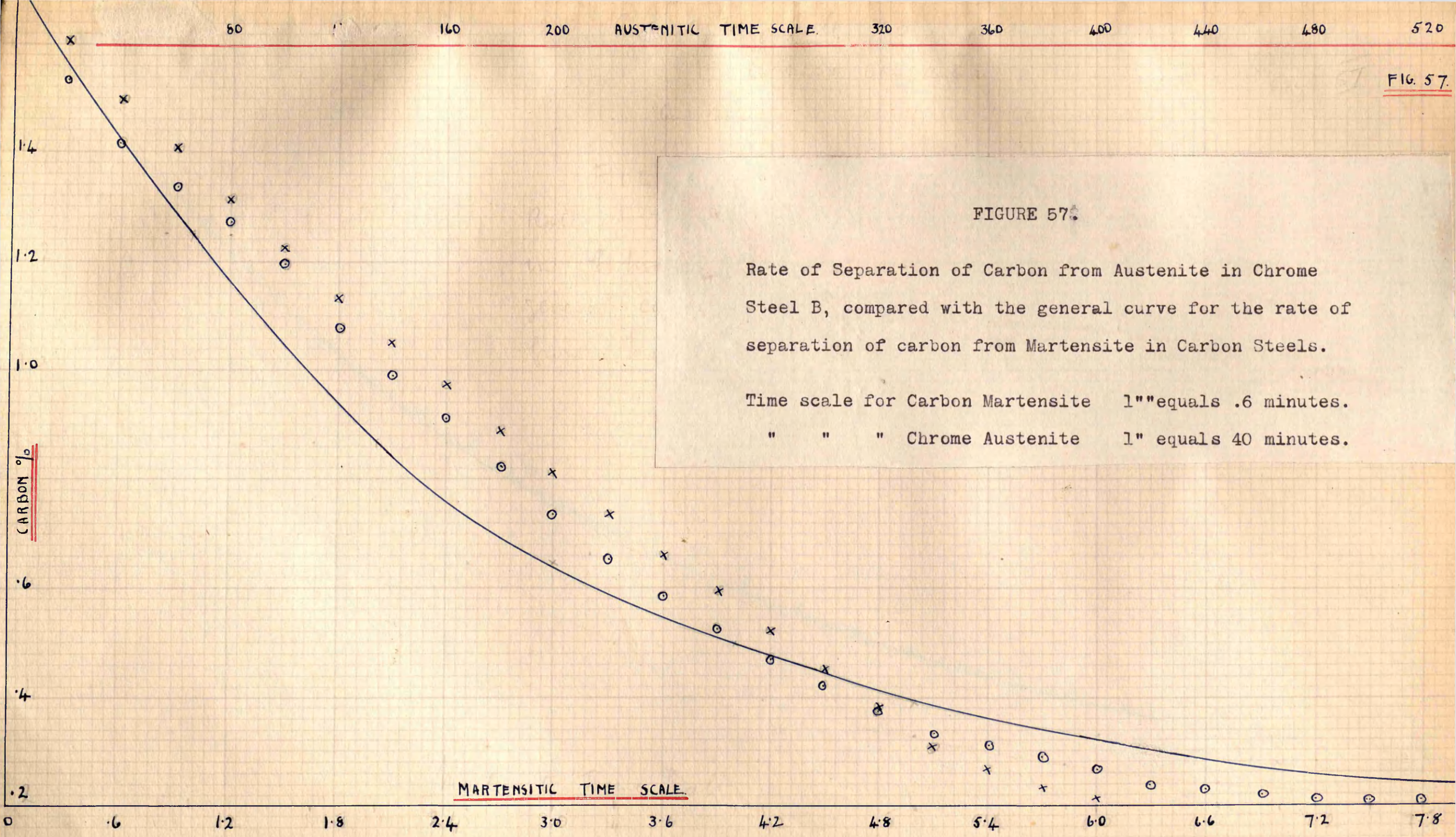


FIGURE 57.

Rate of Separation of Carbon from Austenite in Chrome Steel B, compared with the general curve for the rate of separation of carbon from Martensite in Carbon Steels.

Time scale for Carbon Martensite 1""equals .6 minutes.

" " " Chrome Austenite 1" equals 40 minutes.

FIG. 58.

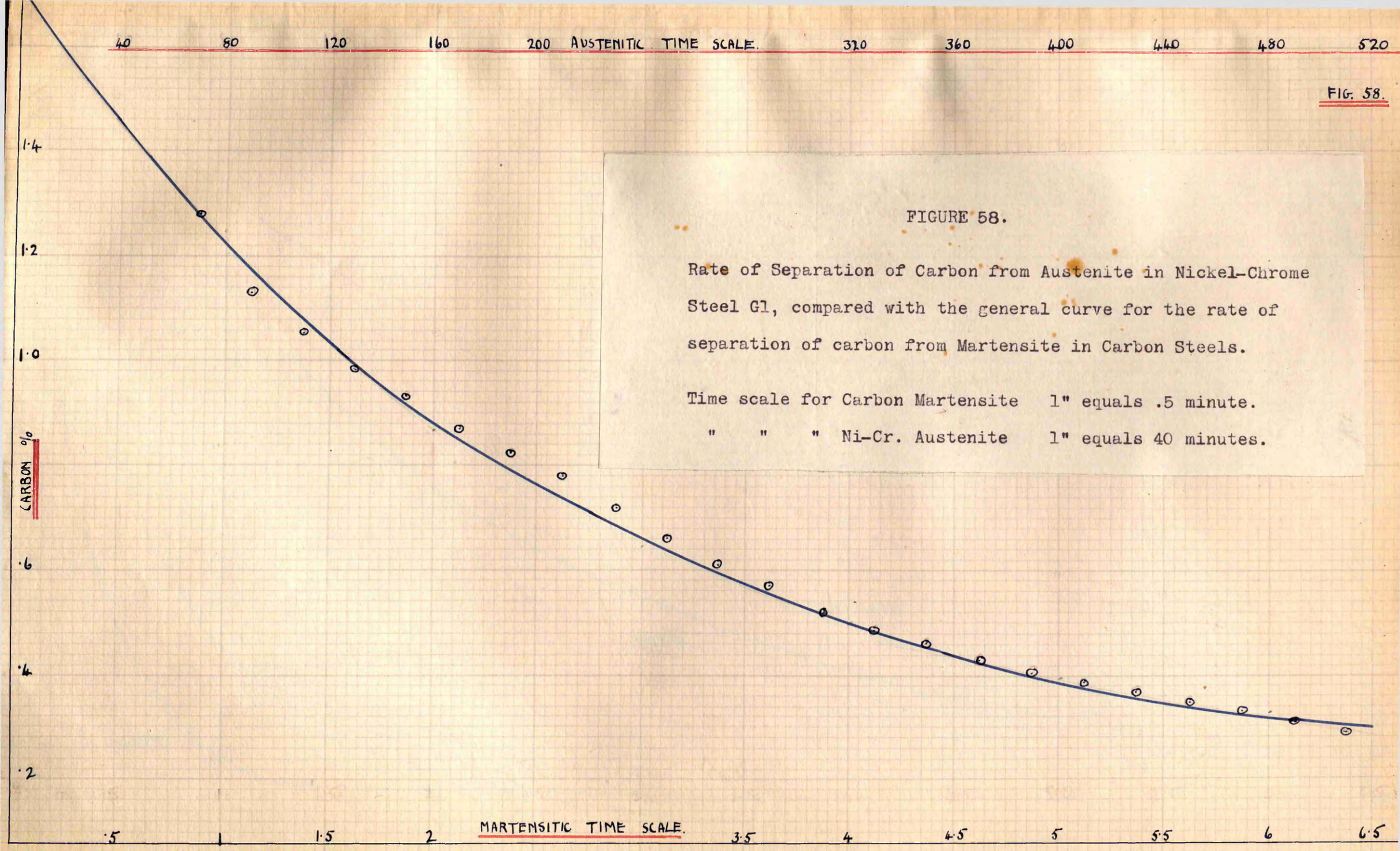


FIGURE 58.

Rate of Separation of Carbon from Austenite in Nickel-Chrome Steel G1, compared with the general curve for the rate of separation of carbon from Martensite in Carbon Steels.

Time scale for Carbon Martensite 1" equals .5 minute.

" " " Ni-Cr. Austenite 1" equals 40 minutes.

40 80 120 160 200 240 280 320 360 400 440 480

TIME IN MINUTES.

FIGURE 59.

COMPARISON OF THE RATE OF SEPARATION OF CARBON FROM SOLUTION IN AUSTENITE IN VARIOUS STEELS.

1·2
1·0
·8
·6
·4
·2

NICKEL-CHROME.

CHROME.

NICKEL.

CARBON.

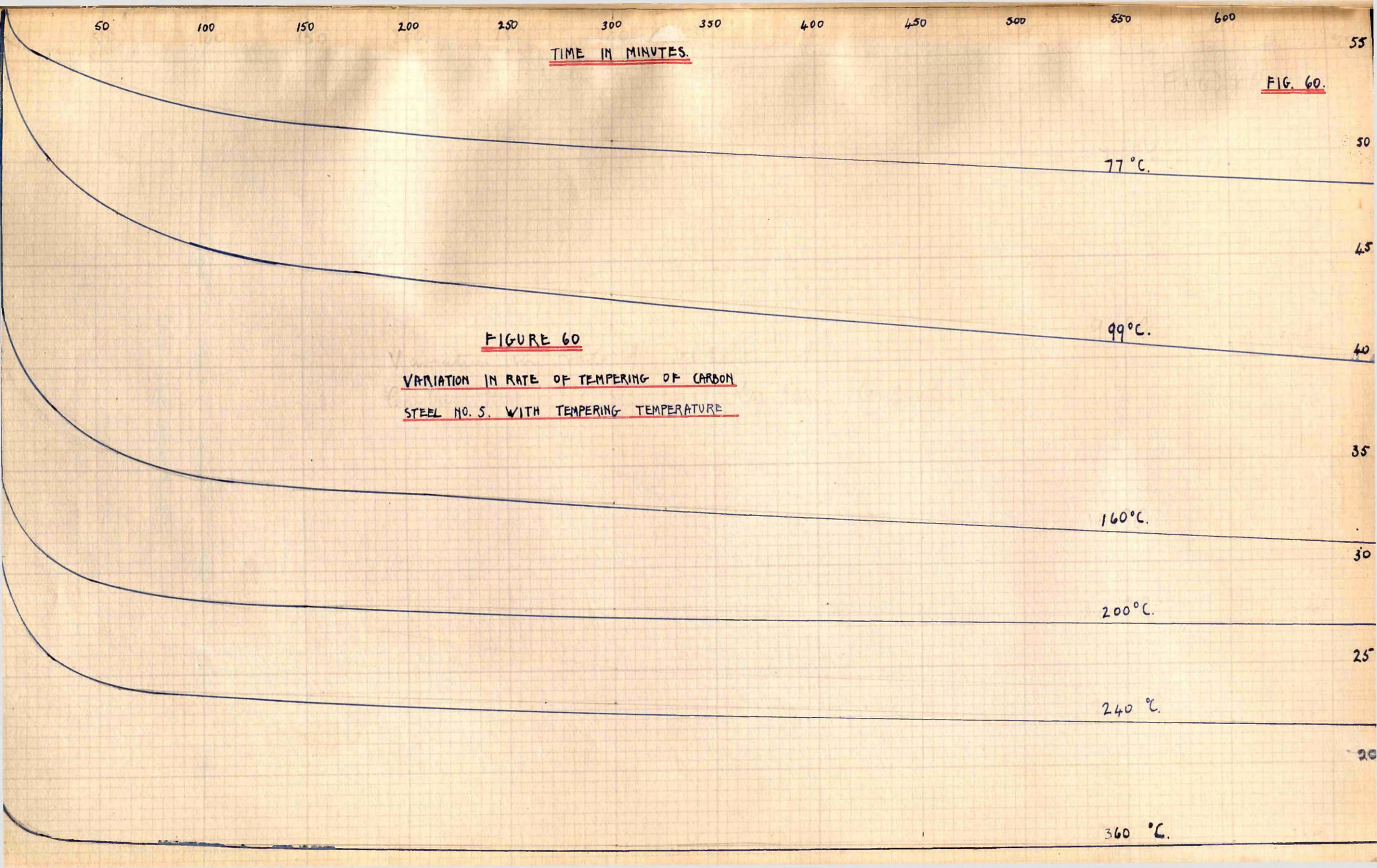


FIG. 60.

FIGURE 60

VARIATION IN RATE OF TEMPERING OF CARBON STEEL NO. 5. WITH TEMPERING TEMPERATURE.

FIGURE 61.

Curve showing the rate of tempering
against tempering temperature
assuming that the rate at 240° C. equals 1.

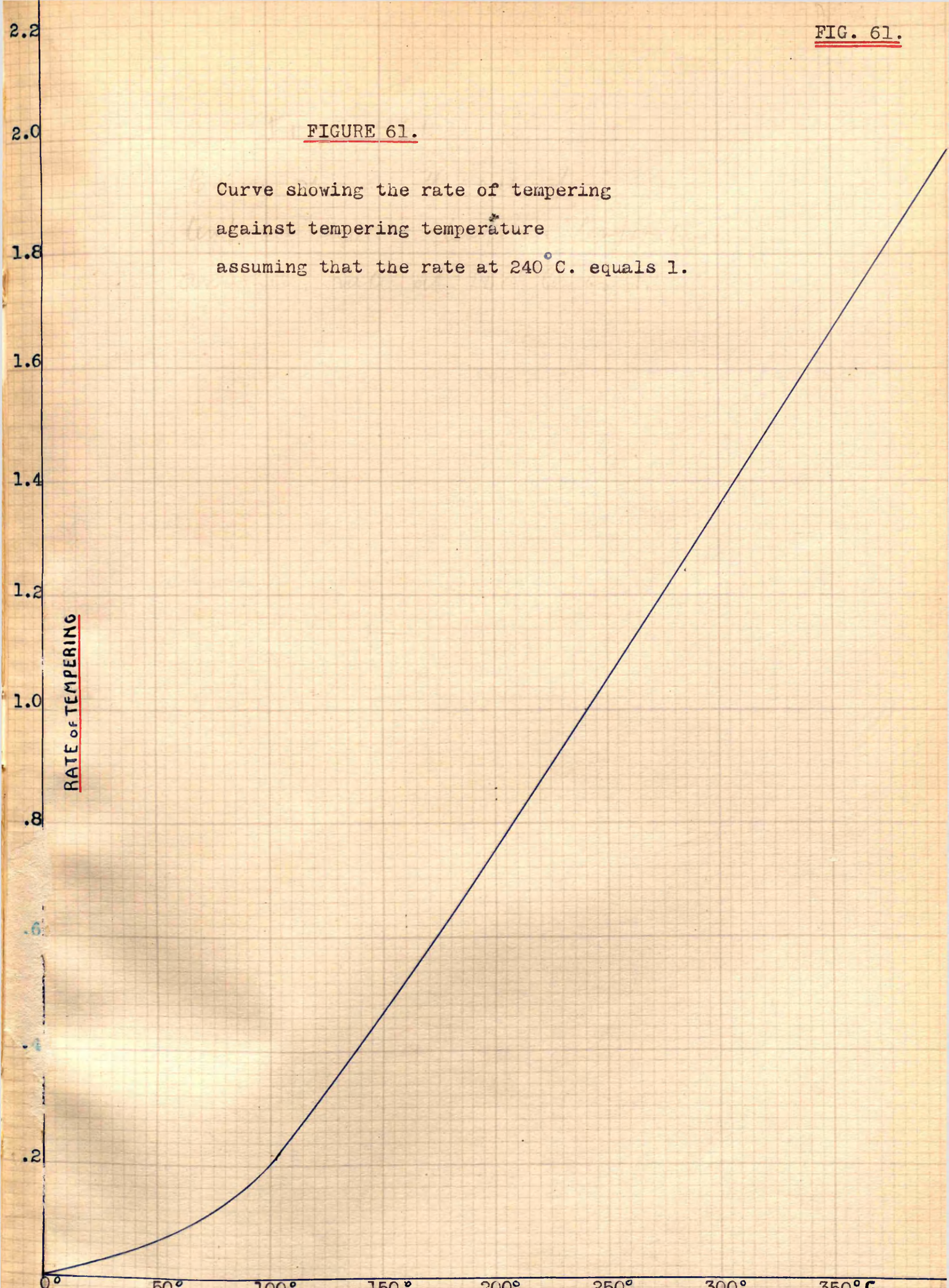


FIG. 62.

FIGURE 62.

Dilatation Curve of Carbon Steel (.22% C.)

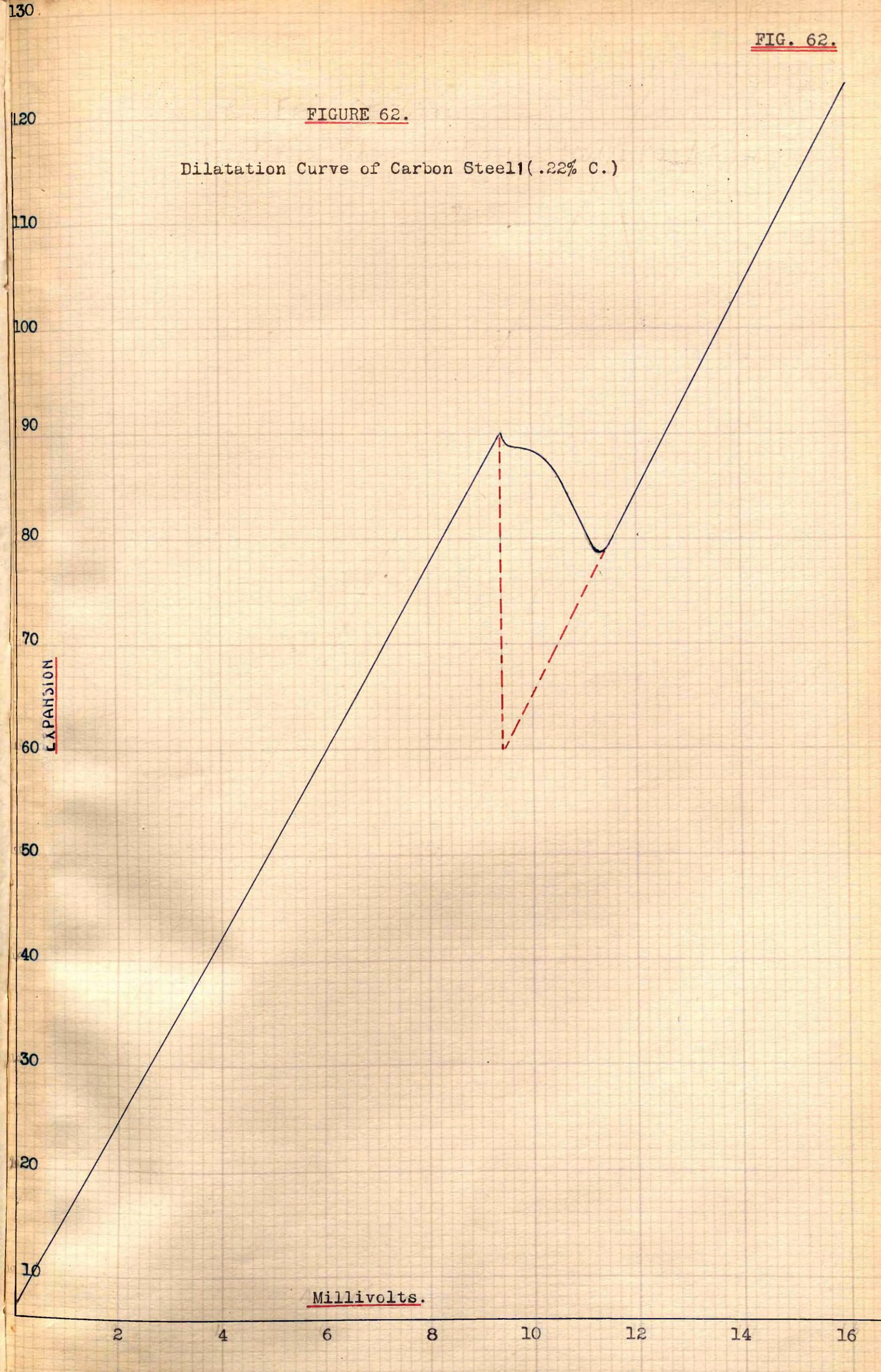


FIG. 63.

FIGURE 63.

Dilatation Curve of Carbon Steel 2 (.48% C.)

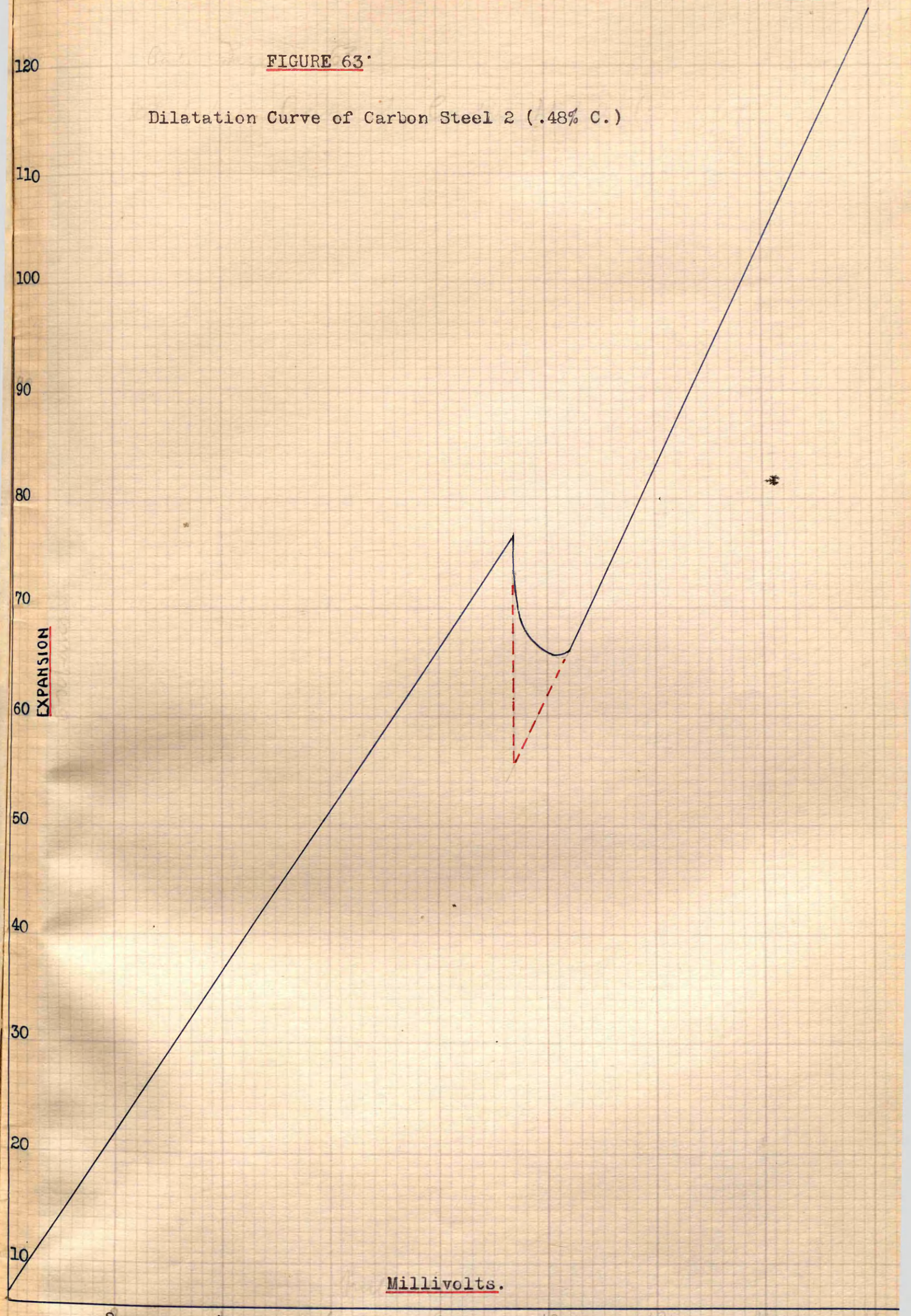


FIG. 64.

FIGURE 64.

Dilatation Curve of Carbon Steel 3 (.71% C.)

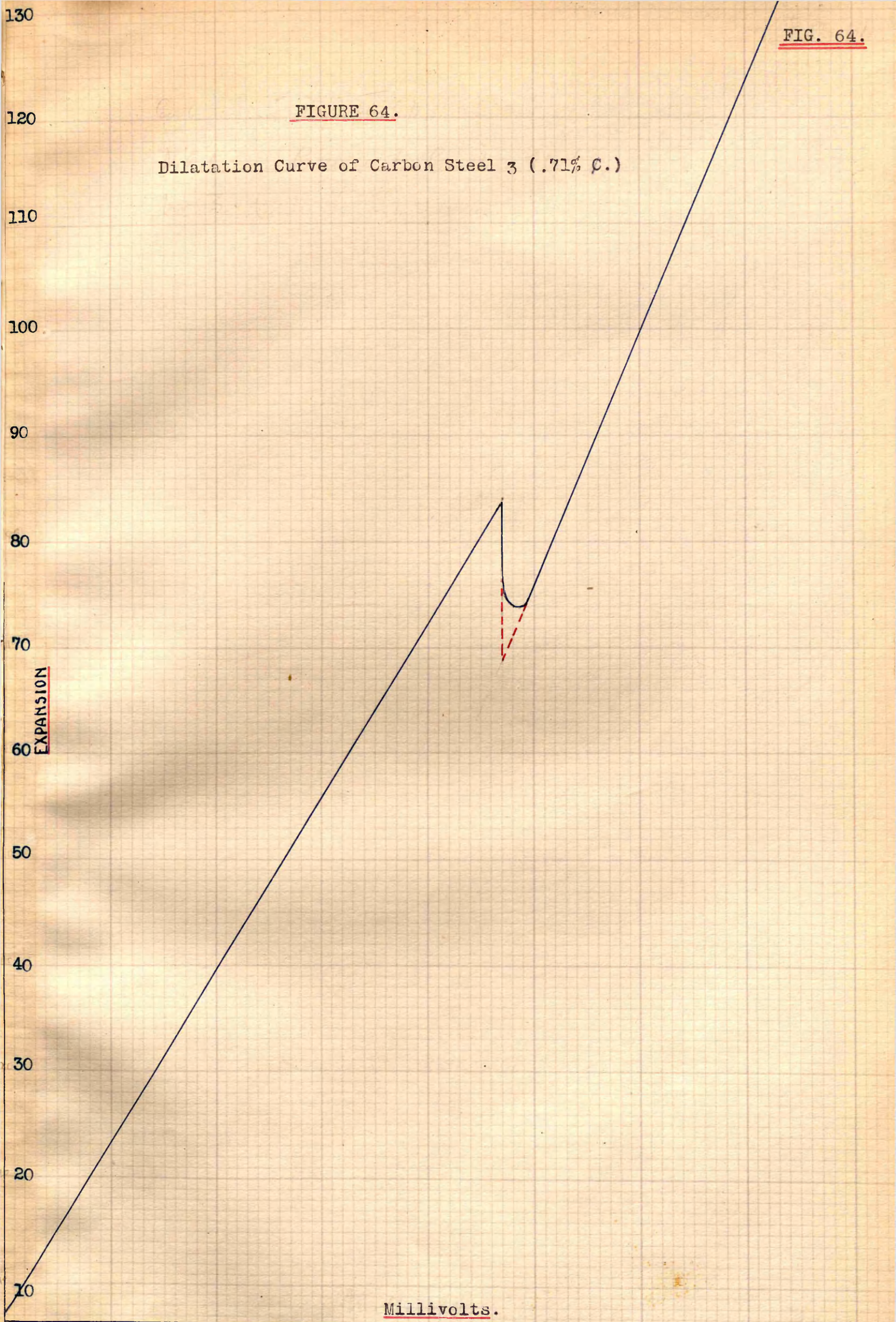


FIG. 65

FIGURE 65.

Dilatation Curve of Carbon Steel 4 (.9% C.)

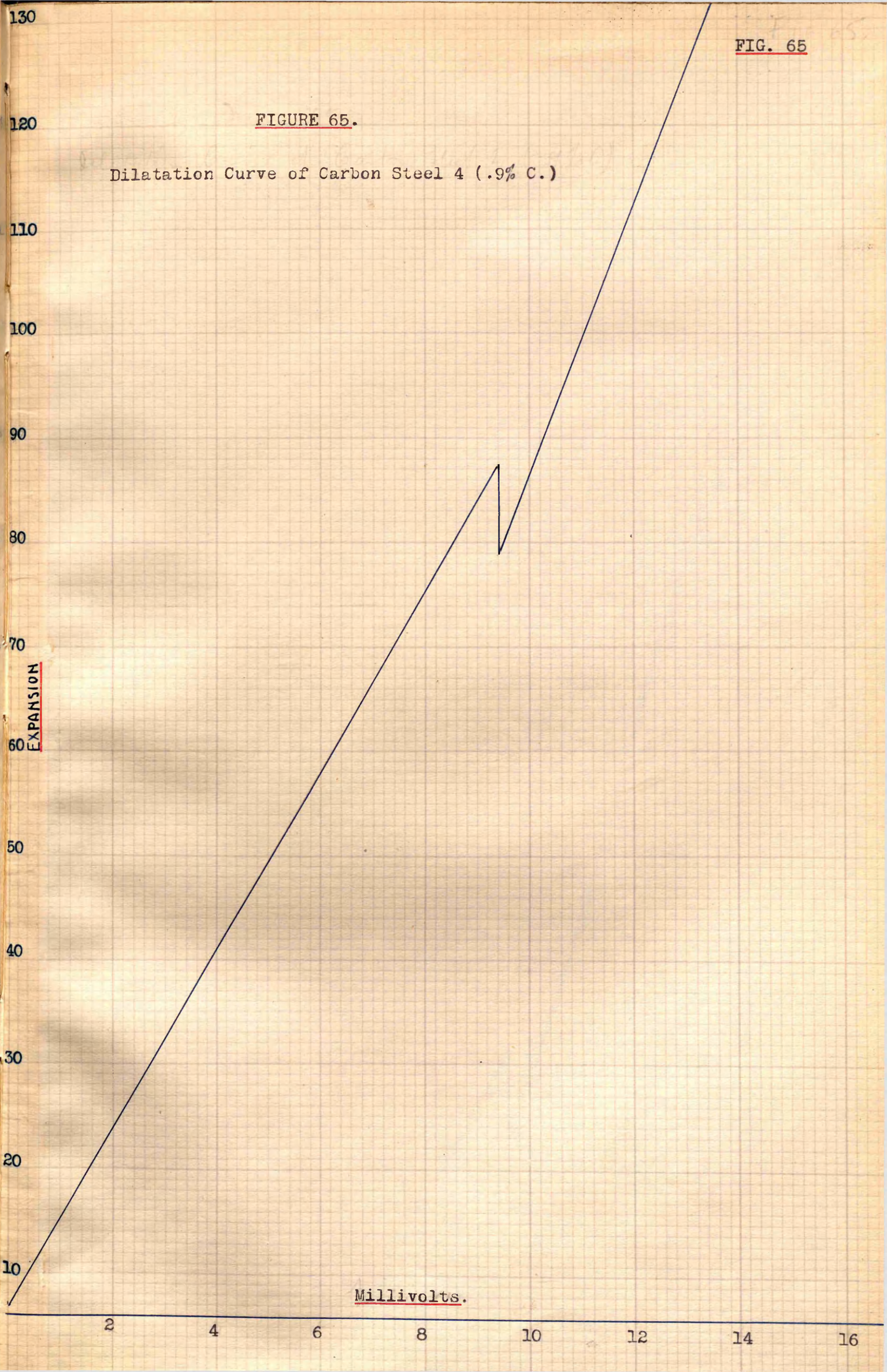


FIG. 66

Figure 66.

Dilatation Curve of Carbon Steel 5 (1.19% C.)

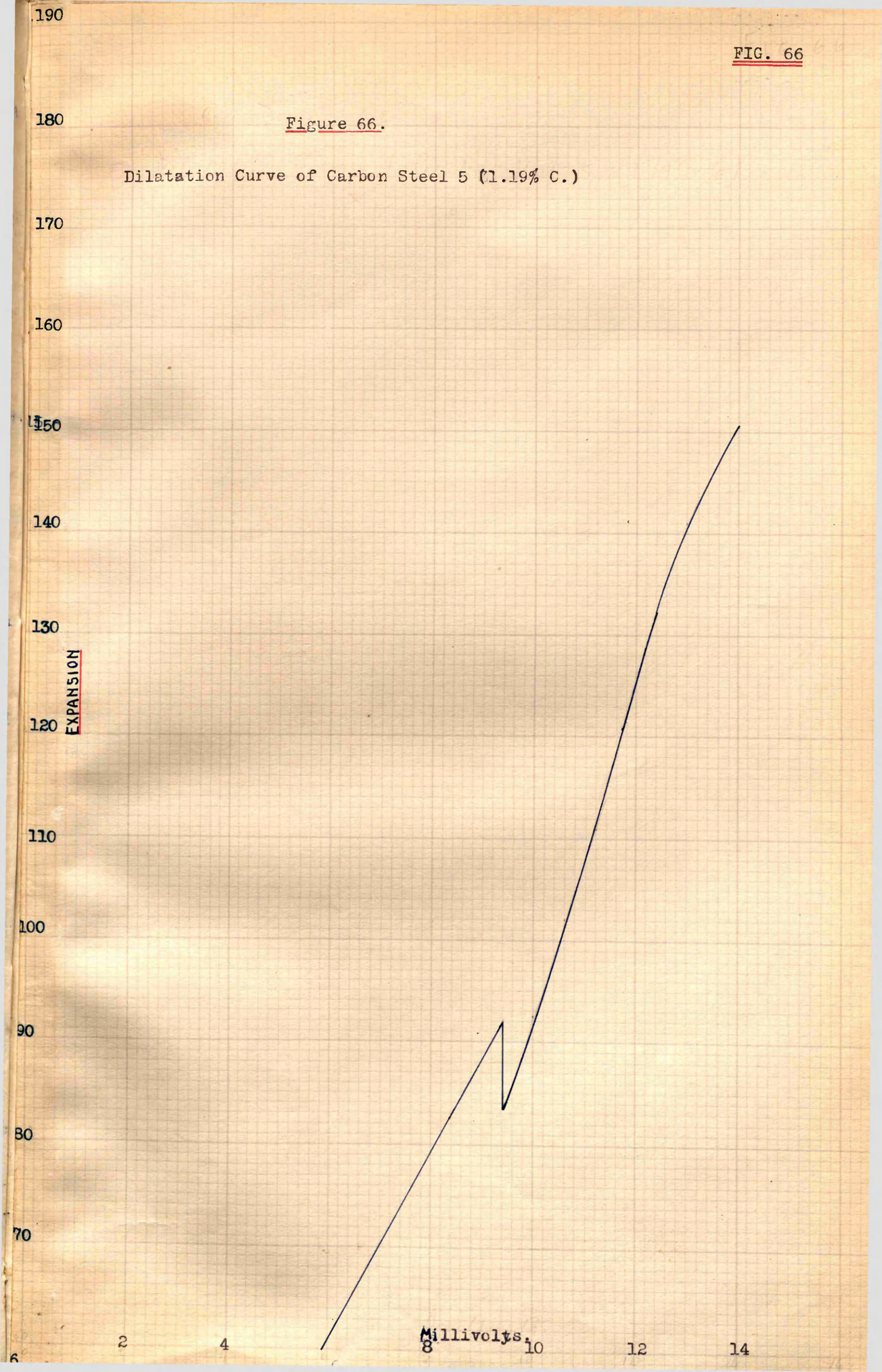


FIG. 67

FIGURE 67.

Dilatation Curve of Carbon Steel 6 (1.61% C.)

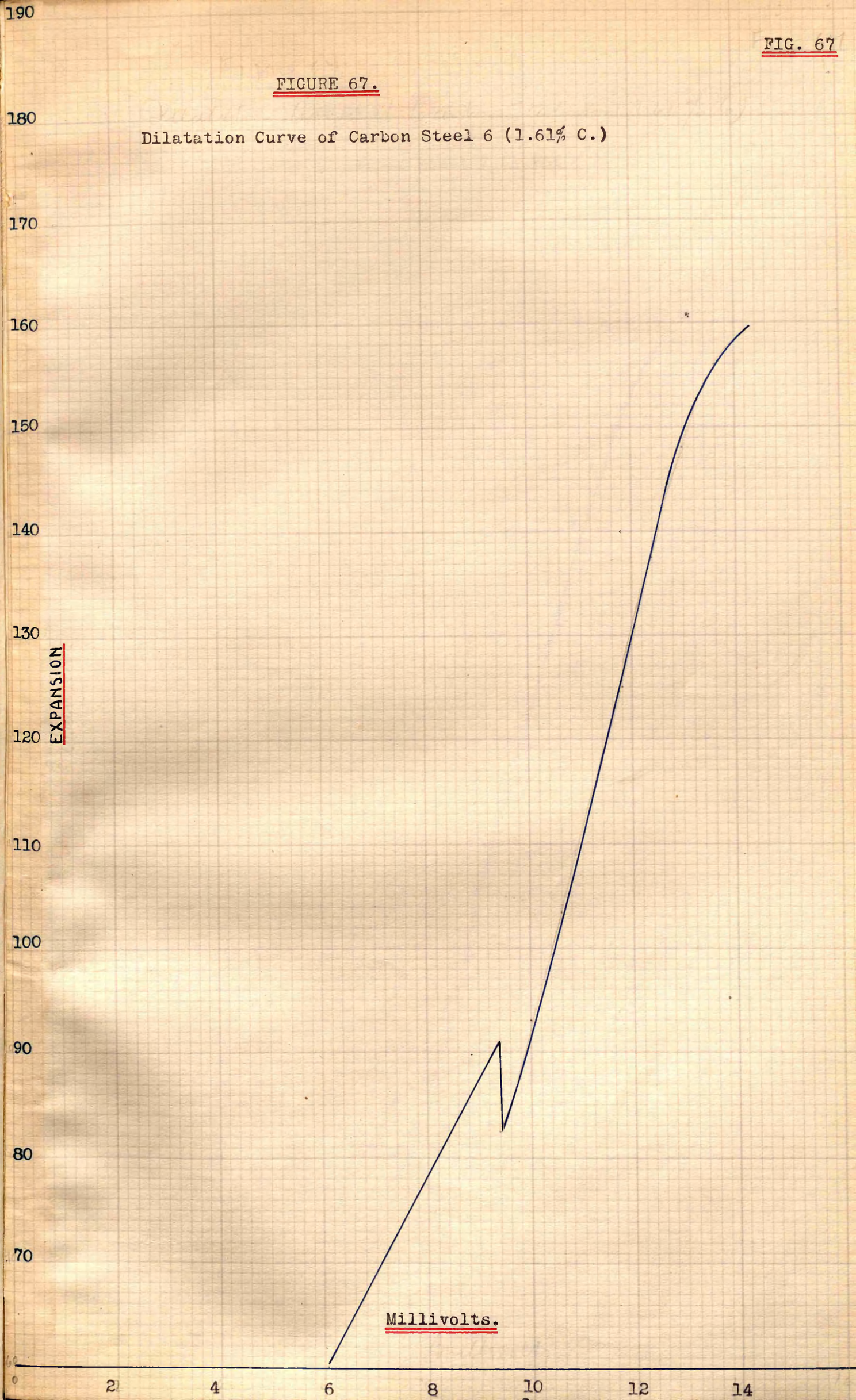


FIGURE 68.

Dilatation Curve of Carbon Steel 7 (1.73% C.)

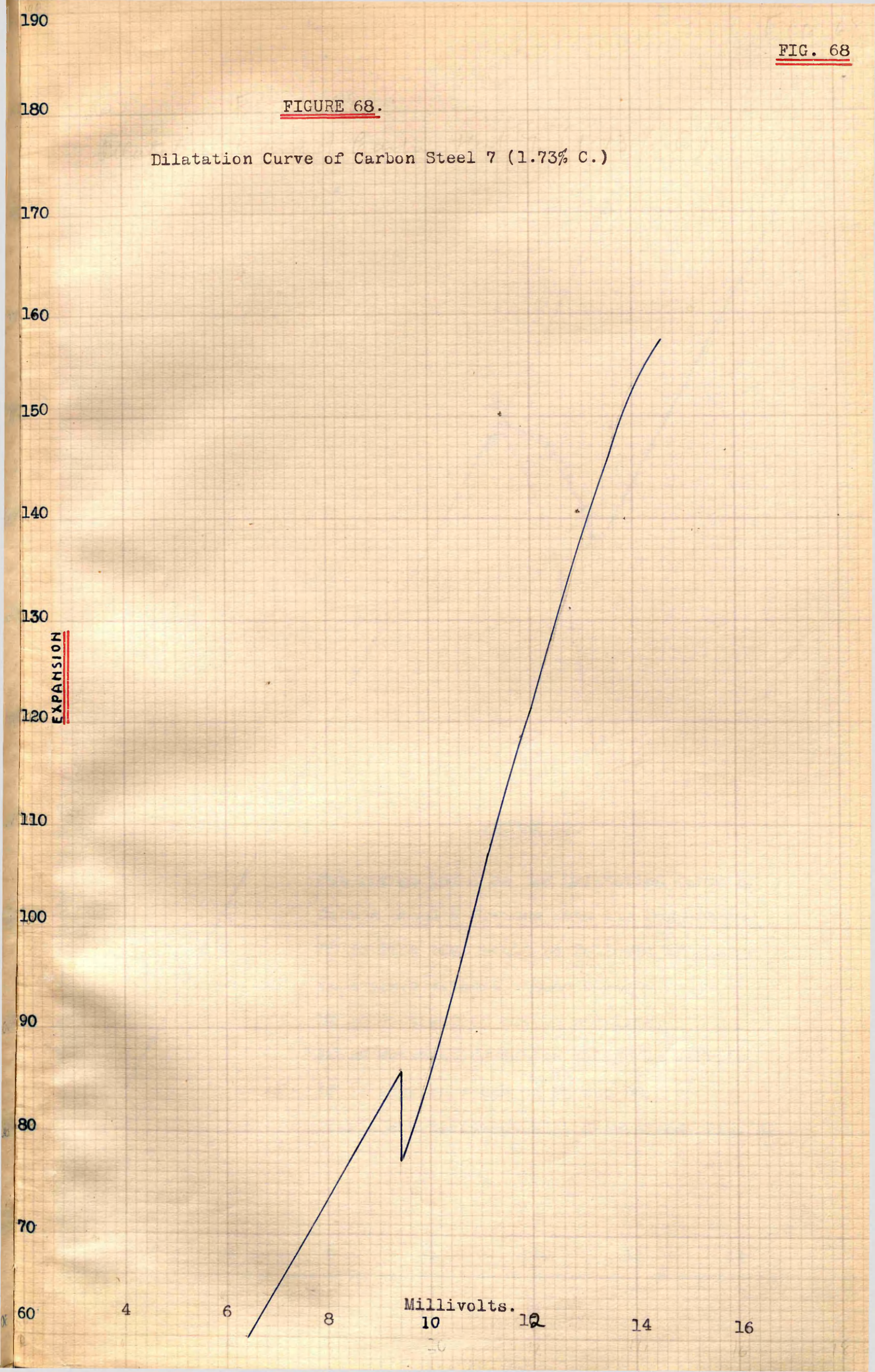


FIG. 69.

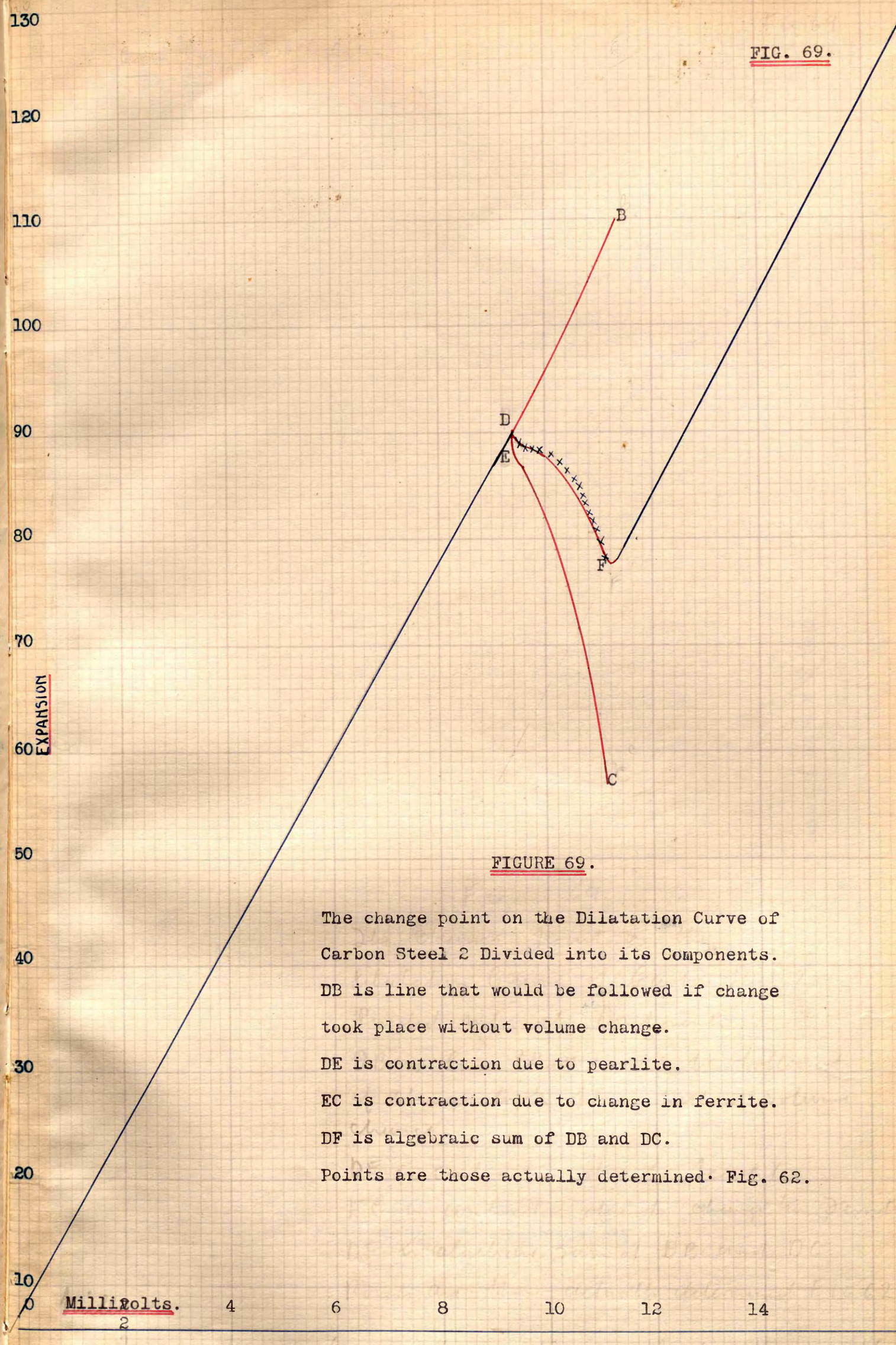


FIGURE 69.

The change point on the Dilatation Curve of Carbon Steel 2 Divided into its Components.

DB is line that would be followed if change took place without volume change.

DE is contraction due to pearlite.

EC is contraction due to change in ferrite.

DF is algebraic sum of DB and DC.

Points are those actually determined. Fig. 62.

FIG 70

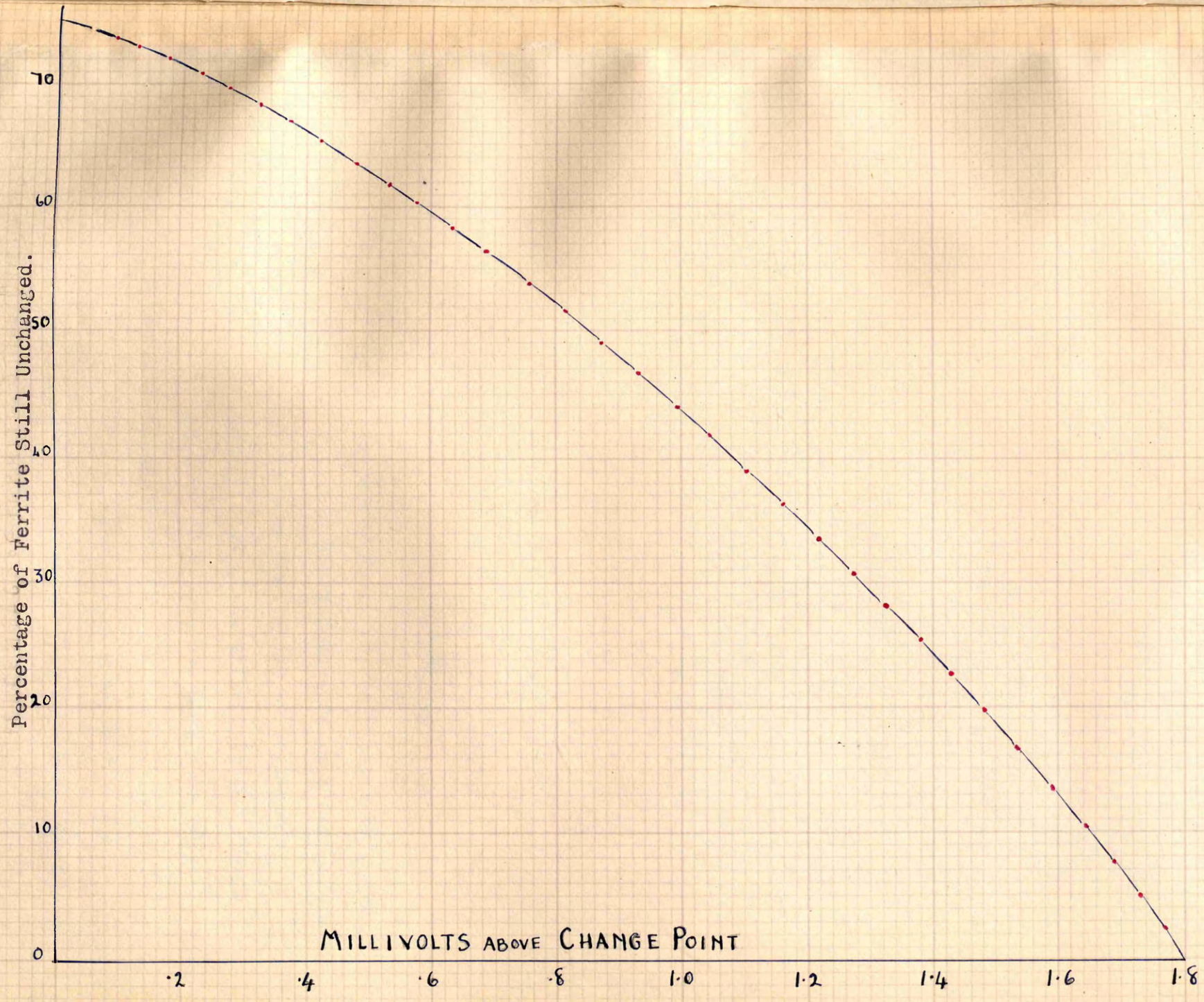

Percentage of Ferrite Still Unchanged.
70
60
50
40
30
20
10
0
MILLIVOLTS ABOVE CHANGE POINT
·2
·4
·6
·8
1·0
1·2
1·4
1·6
1·8

FIG. 71

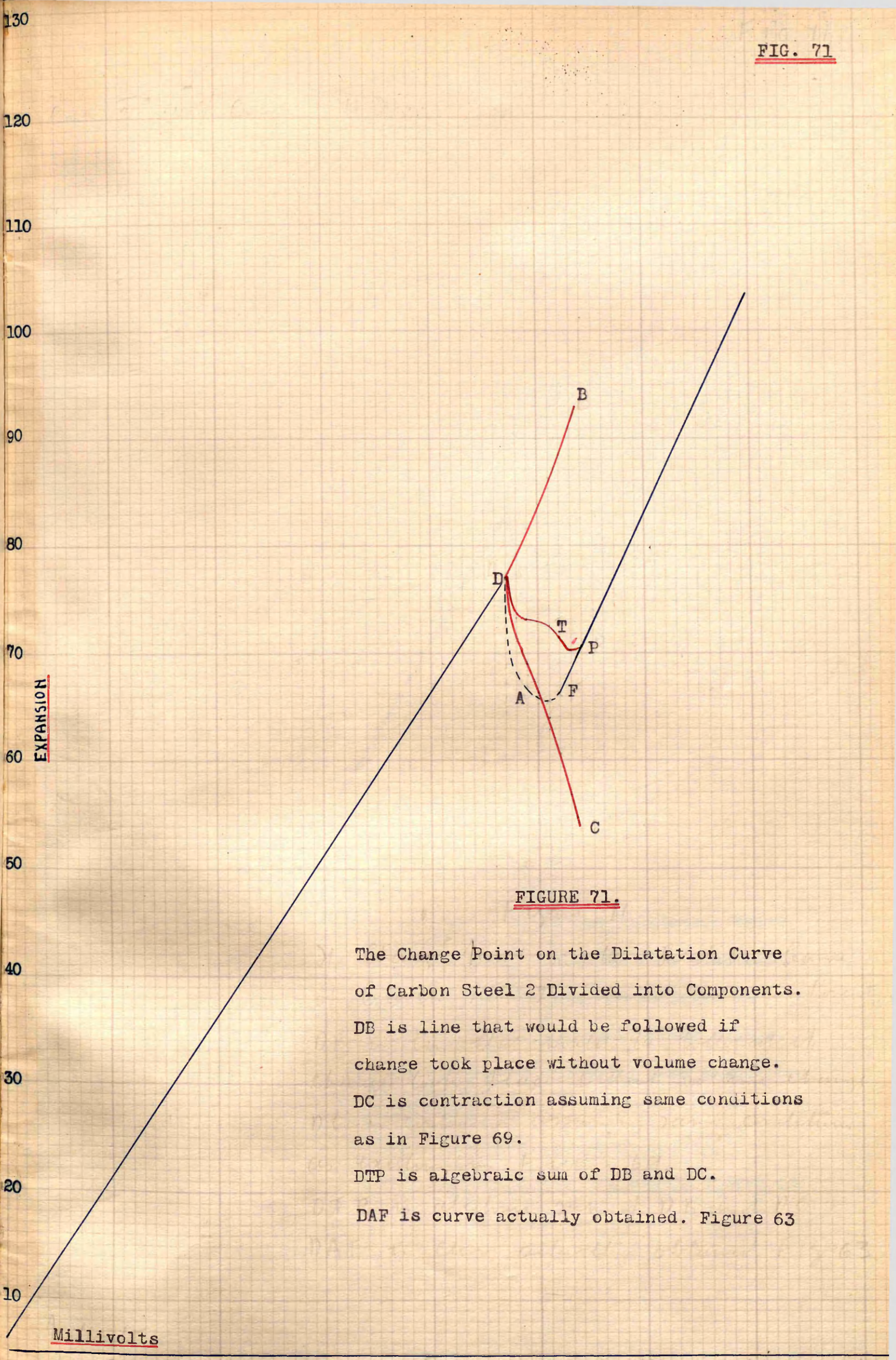


FIGURE 71.

The Change Point on the Dilatation Curve of Carbon Steel 2 Divided into Components.
DB is line that would be followed if change took place without volume change.
DC is contraction assuming same conditions as in Figure 69.
DTP is algebraic sum of DB and DC.
DAF is curve actually obtained. Figure 63

FIG. 72.

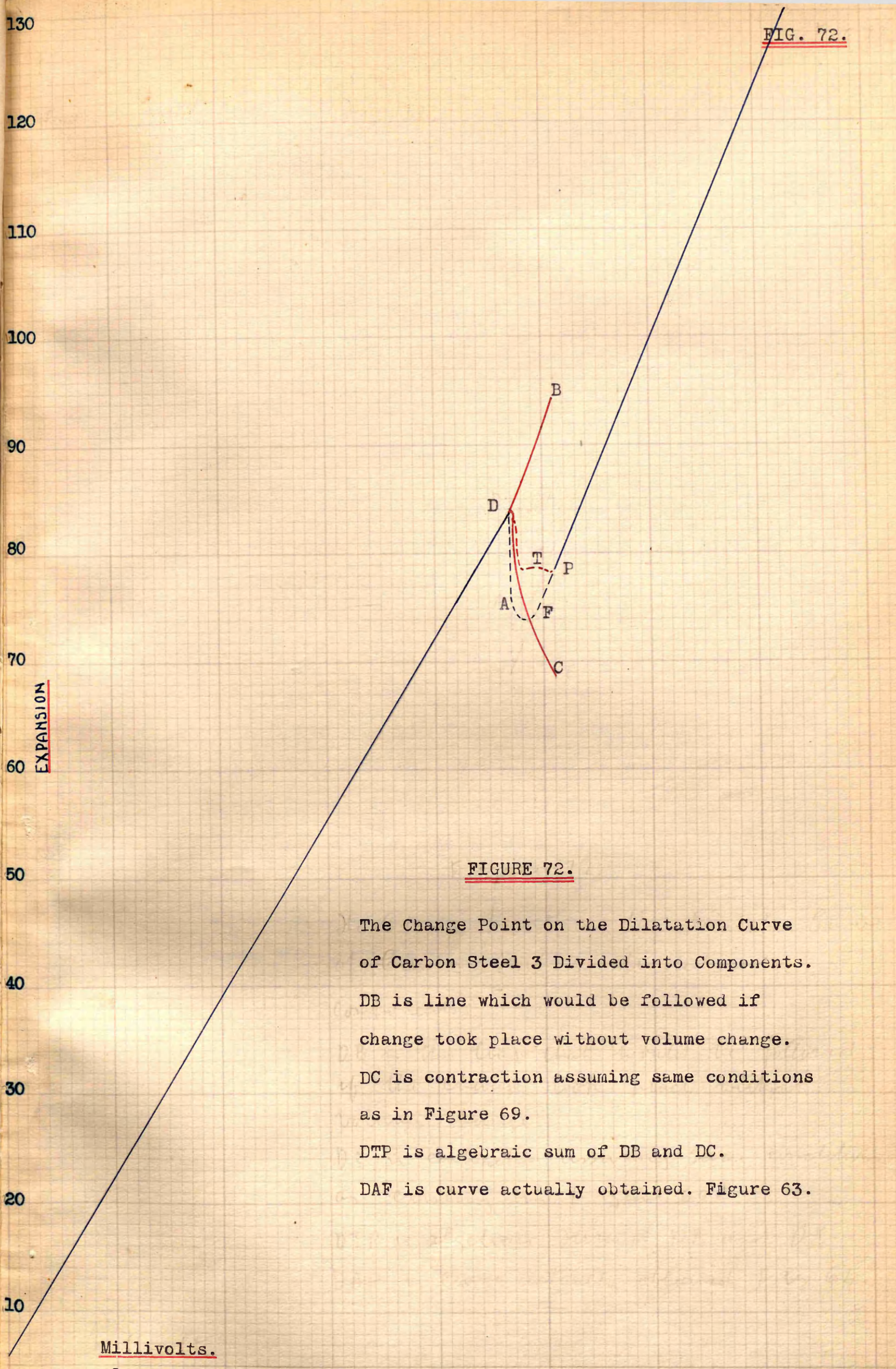


FIGURE 72.

The Change Point on the Dilatation Curve of Carbon Steel 3 Divided into Components.
DB is line which would be followed if change took place without volume change.
DC is contraction assuming same conditions as in Figure 69.
DTP is algebraic sum of DB and DC.
DAF is curve actually obtained. Figure 63.

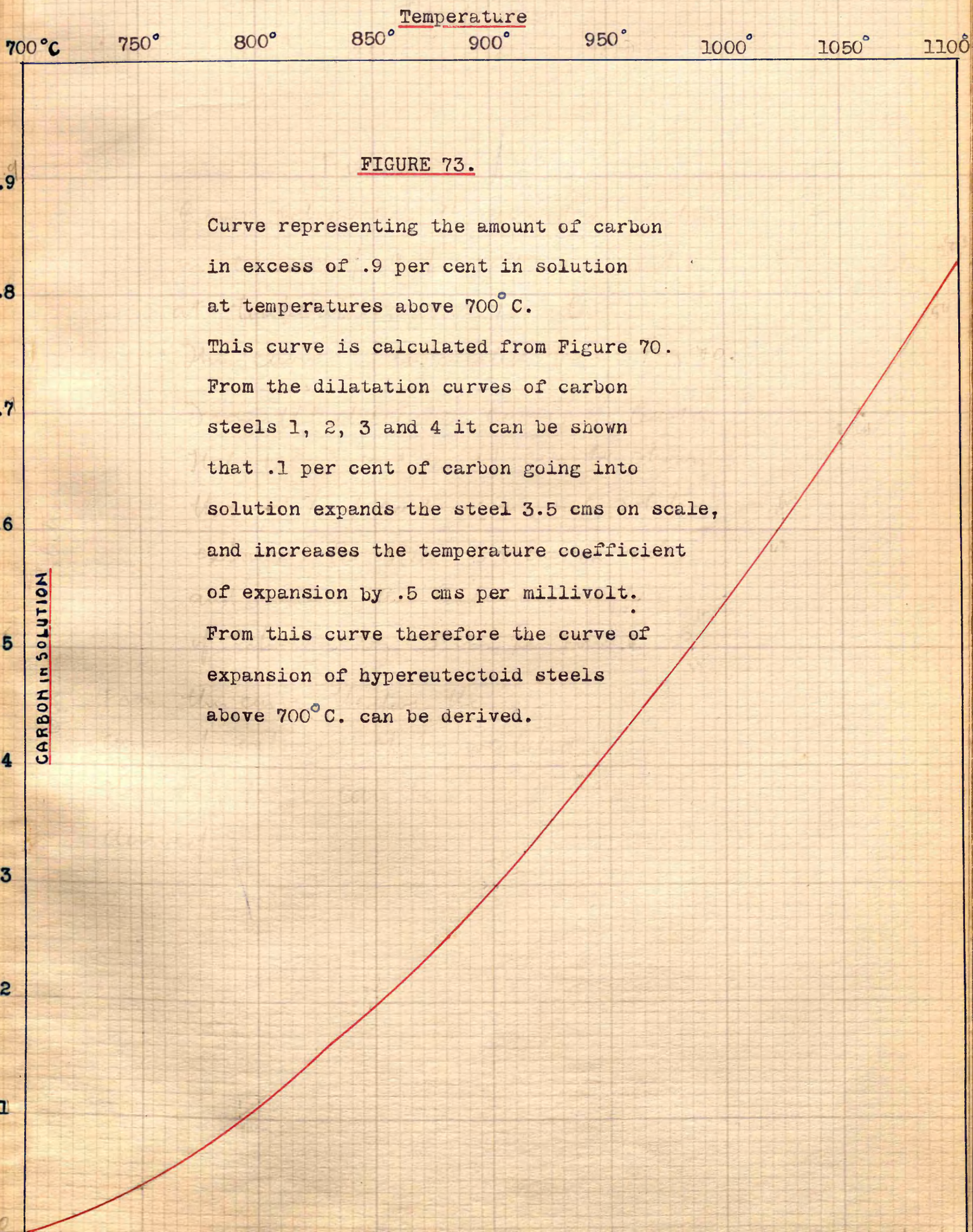


FIGURE 73.

Curve representing the amount of carbon in excess of .9 per cent in solution at temperatures above 700° C.

This curve is calculated from Figure 70.

From the dilatation curves of carbon steels 1, 2, 3 and 4 it can be shown that .1 per cent of carbon going into solution expands the steel 3.5 cms on scale, and increases the temperature coefficient of expansion by .5 cms per millivolt.

From this curve therefore the curve of expansion of hypereutectoid steels above 700° C. can be derived.

FIG. 74.

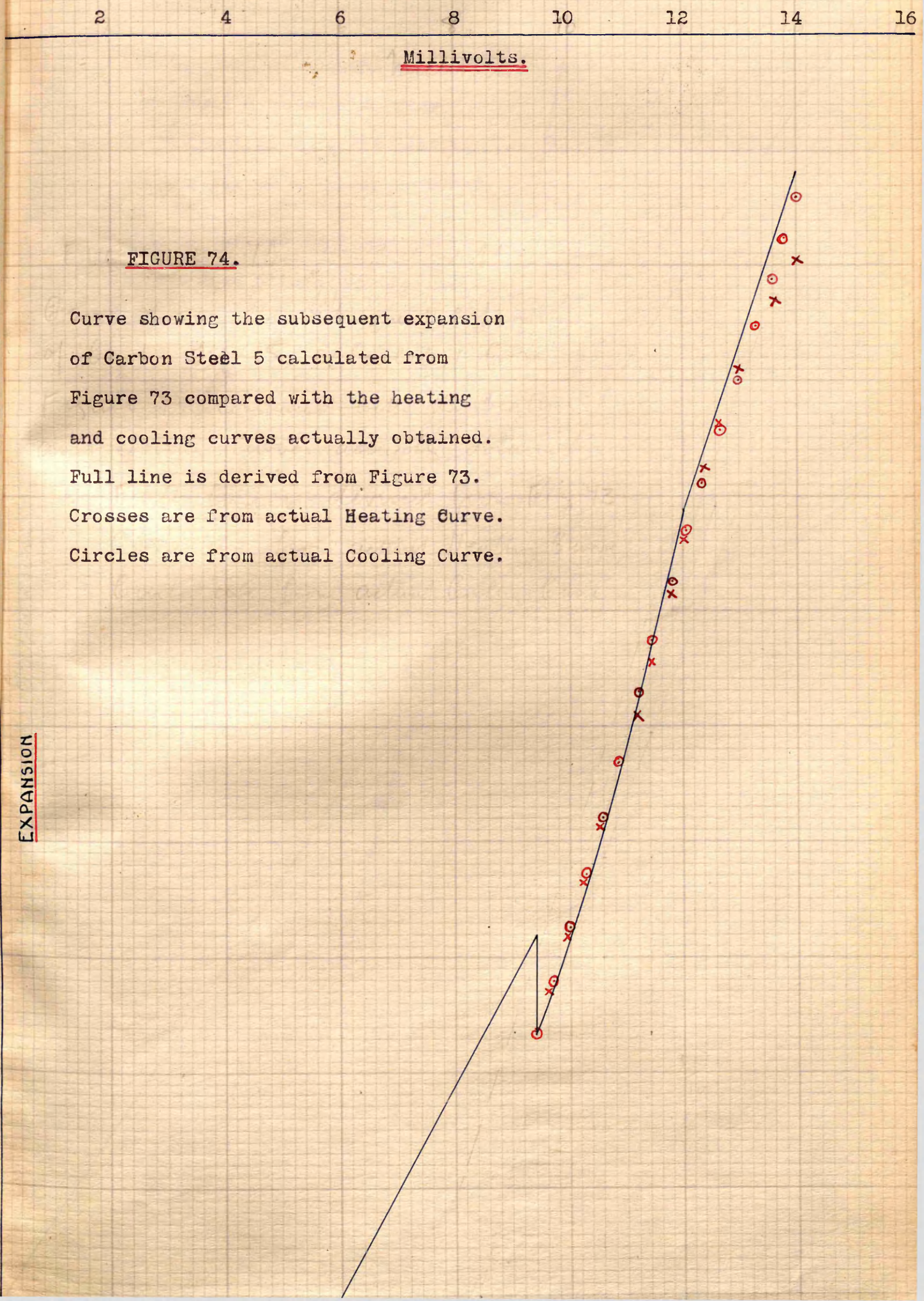


FIGURE 74.

Curve showing the subsequent expansion of Carbon Steel 5 calculated from Figure 73 compared with the heating and cooling curves actually obtained.
Full line is derived from Figure 73.
Crosses are from actual Heating Curve.
Circles are from actual Cooling Curve.

FIGURE 75.

Curve showing the subsequent expansion of Carbon Steel 6 calculated from Figure 73 compared with the heating and cooling curves actually obtained.

Full line is derived from Figure 73.
Crosses are from actual Heating Curve.
Circles are from actual Cooling Curve.

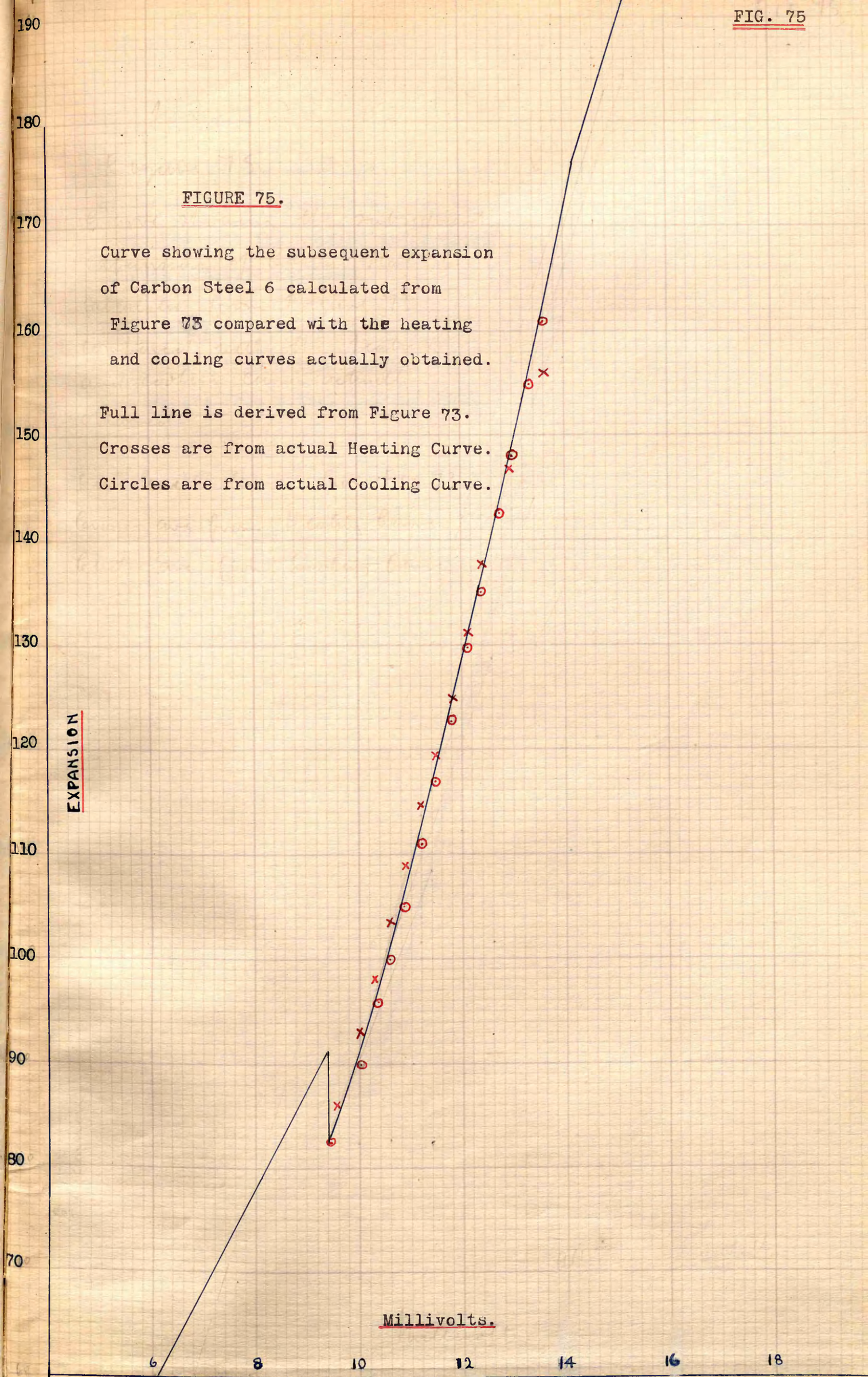


Figure 76.

Curve showing the subsequent expansion of Carbon Steel 7 calculated from Figure 73 compared with the heating and cooling curves actually obtained.

Full line is derived from Figure 73.
Crosses are from actual Heating Curve.
Circles are from actual Cooling Curve.

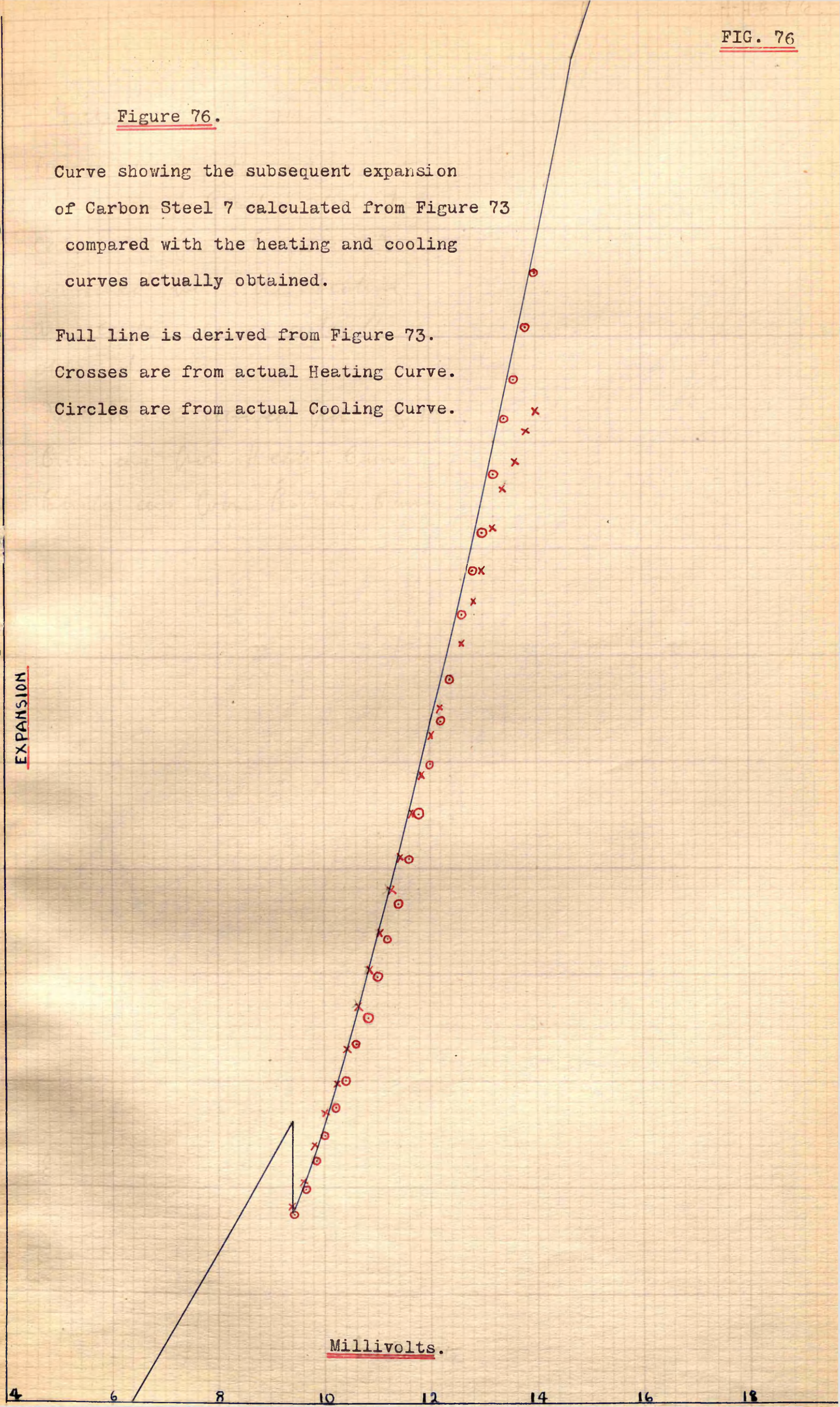


FIG. 77

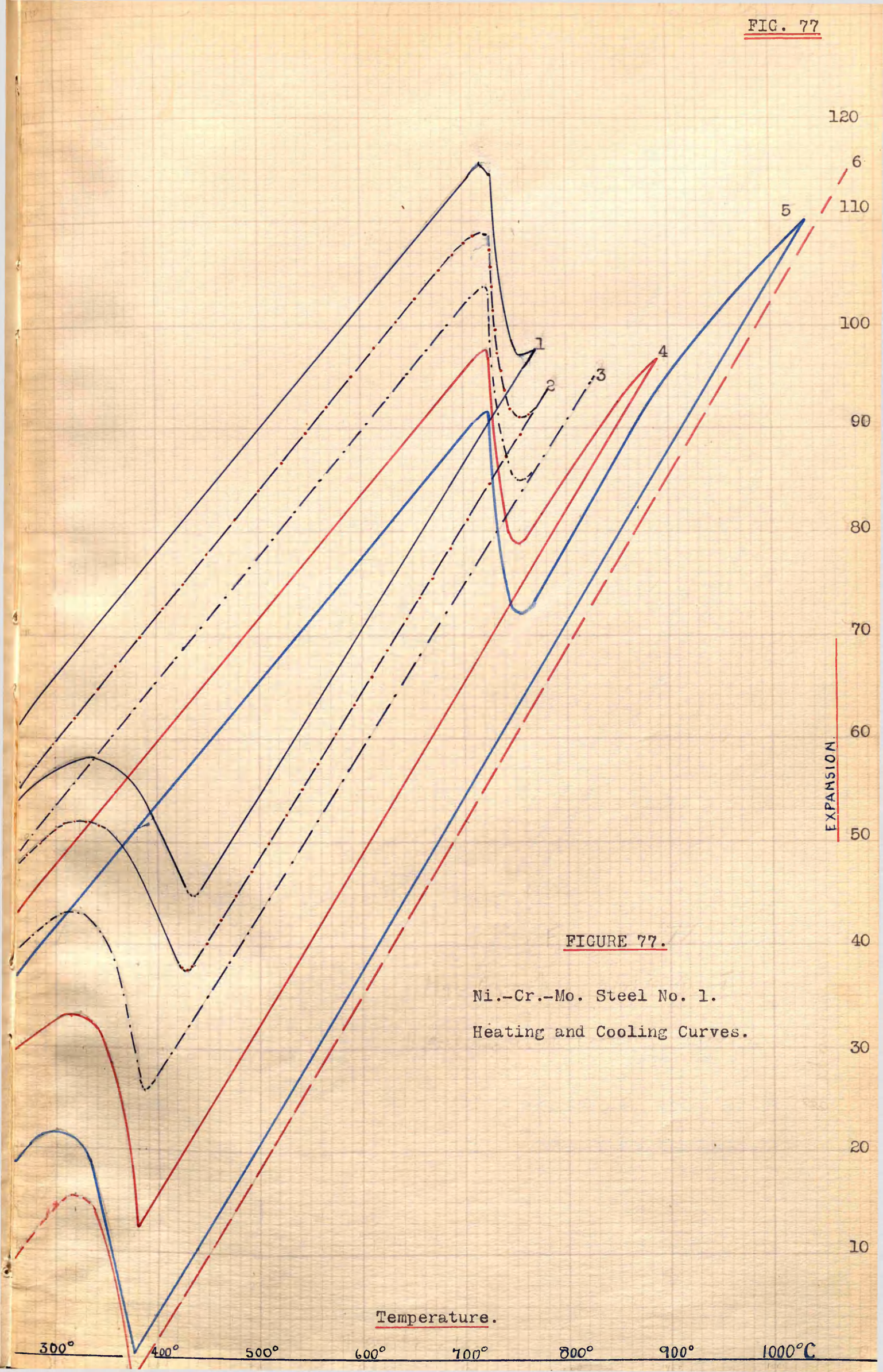


FIGURE 77.

Ni.-Cr.-Mo. Steel No. 1.

Heating and Cooling Curves.

FIG. 78

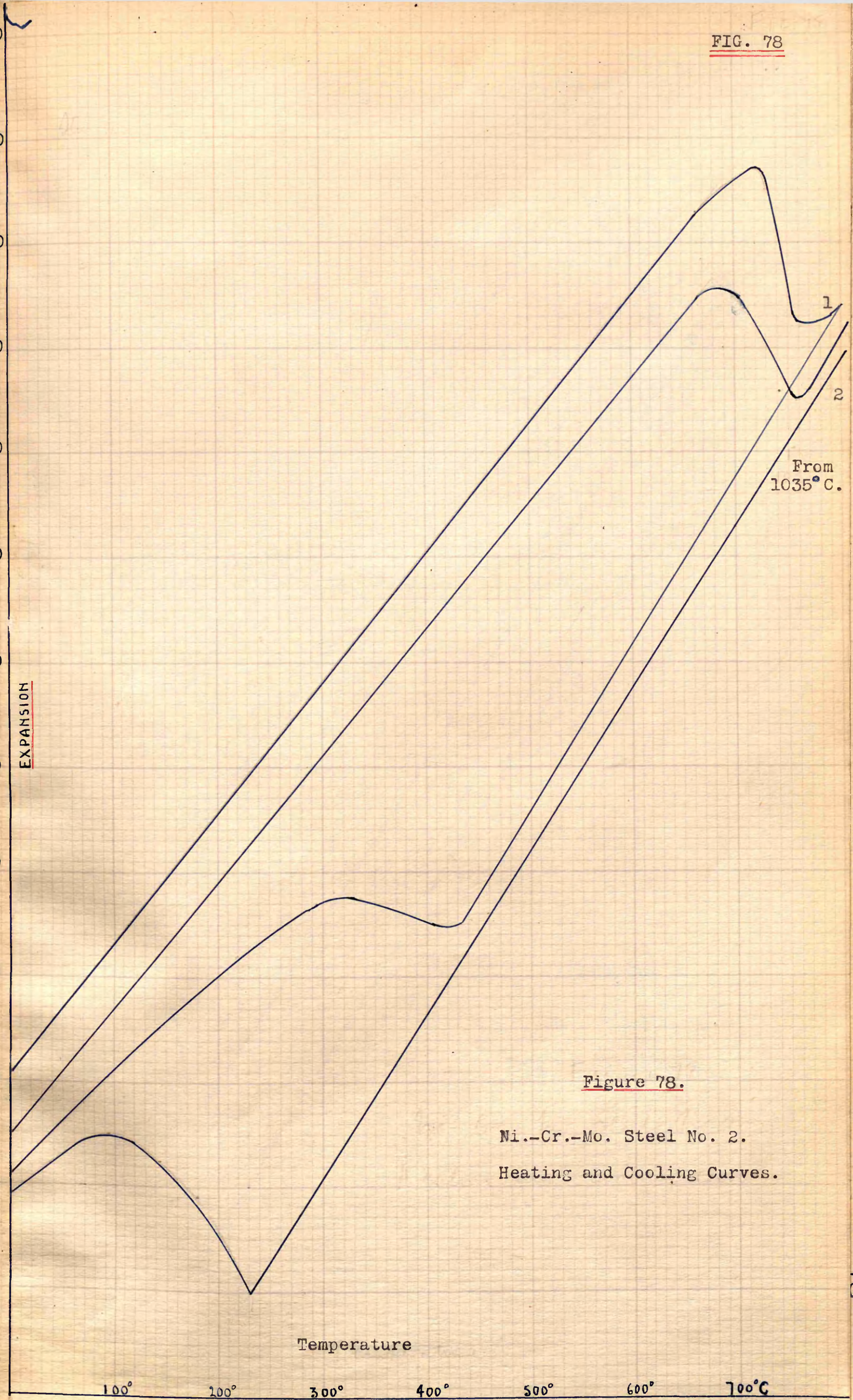


Figure 78.

Ni.-Cr.-Mo. Steel No. 2.

Heating and Cooling Curves.

FIG. 79

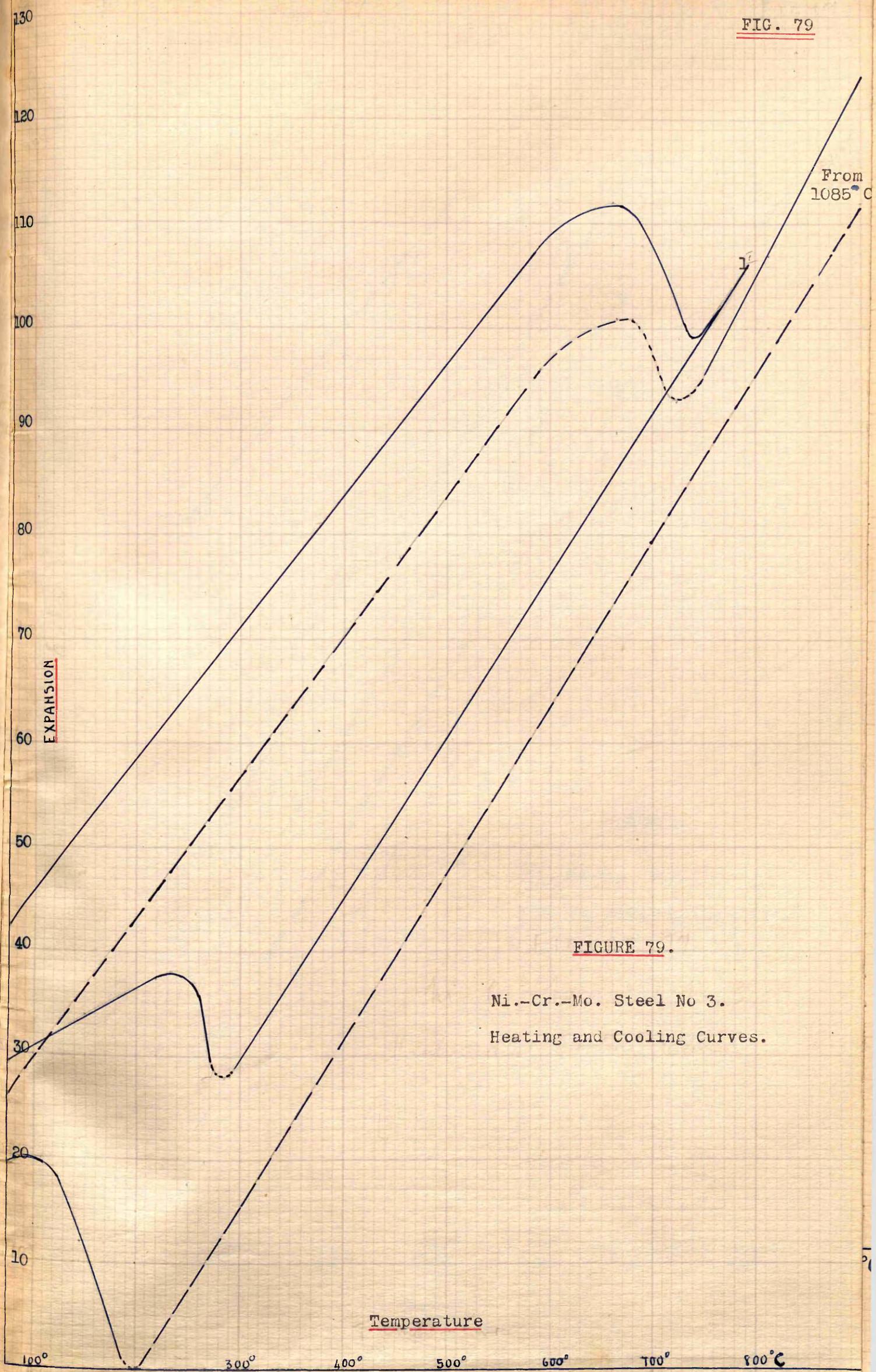


FIGURE 79.

Ni.-Cr.-Mo. Steel No 3.

Heating and Cooling Curves.

FIG. 80

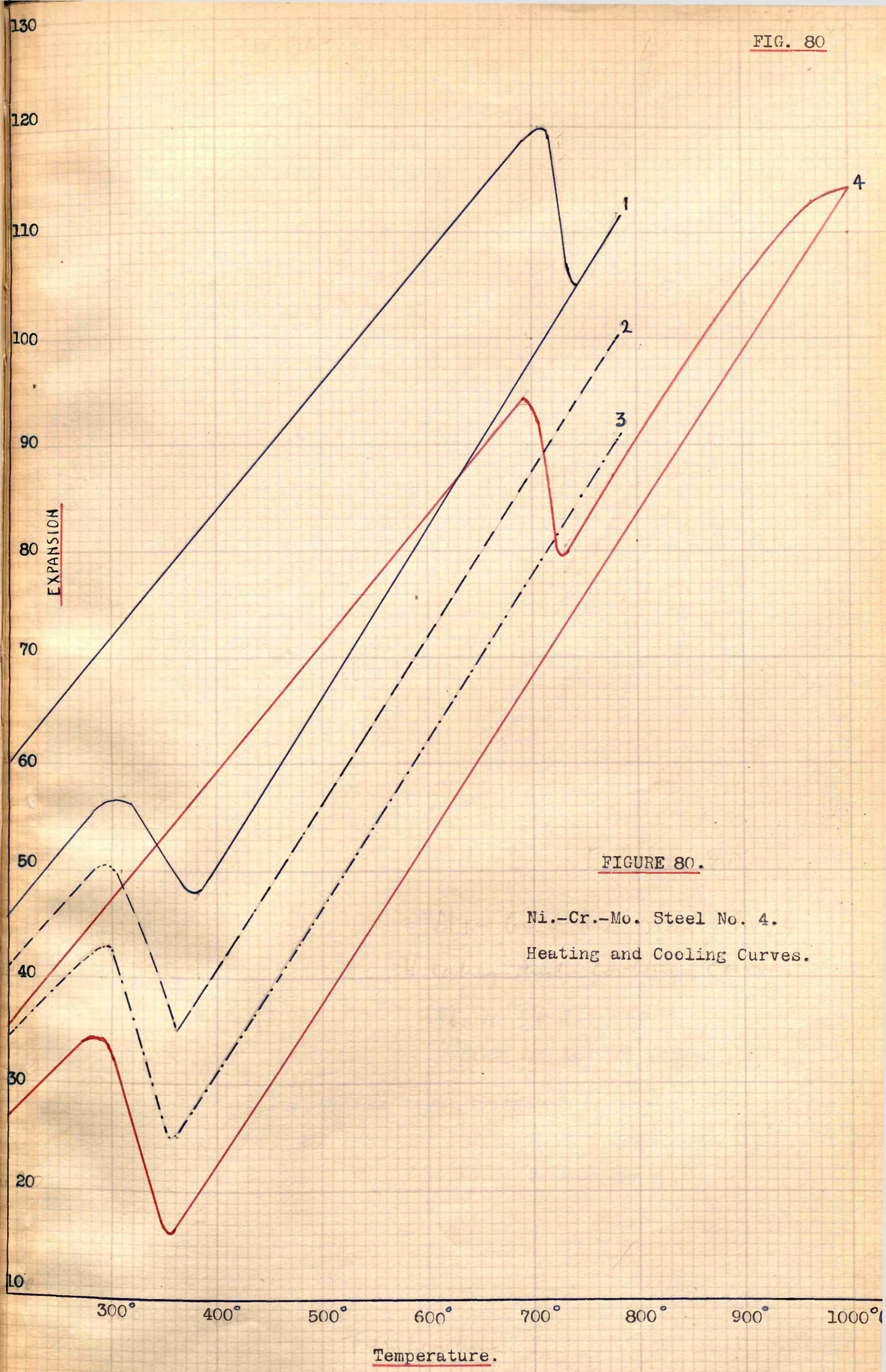


FIGURE 80.

Ni.-Cr.-Mo. Steel No. 4.

Heating and Cooling Curves.

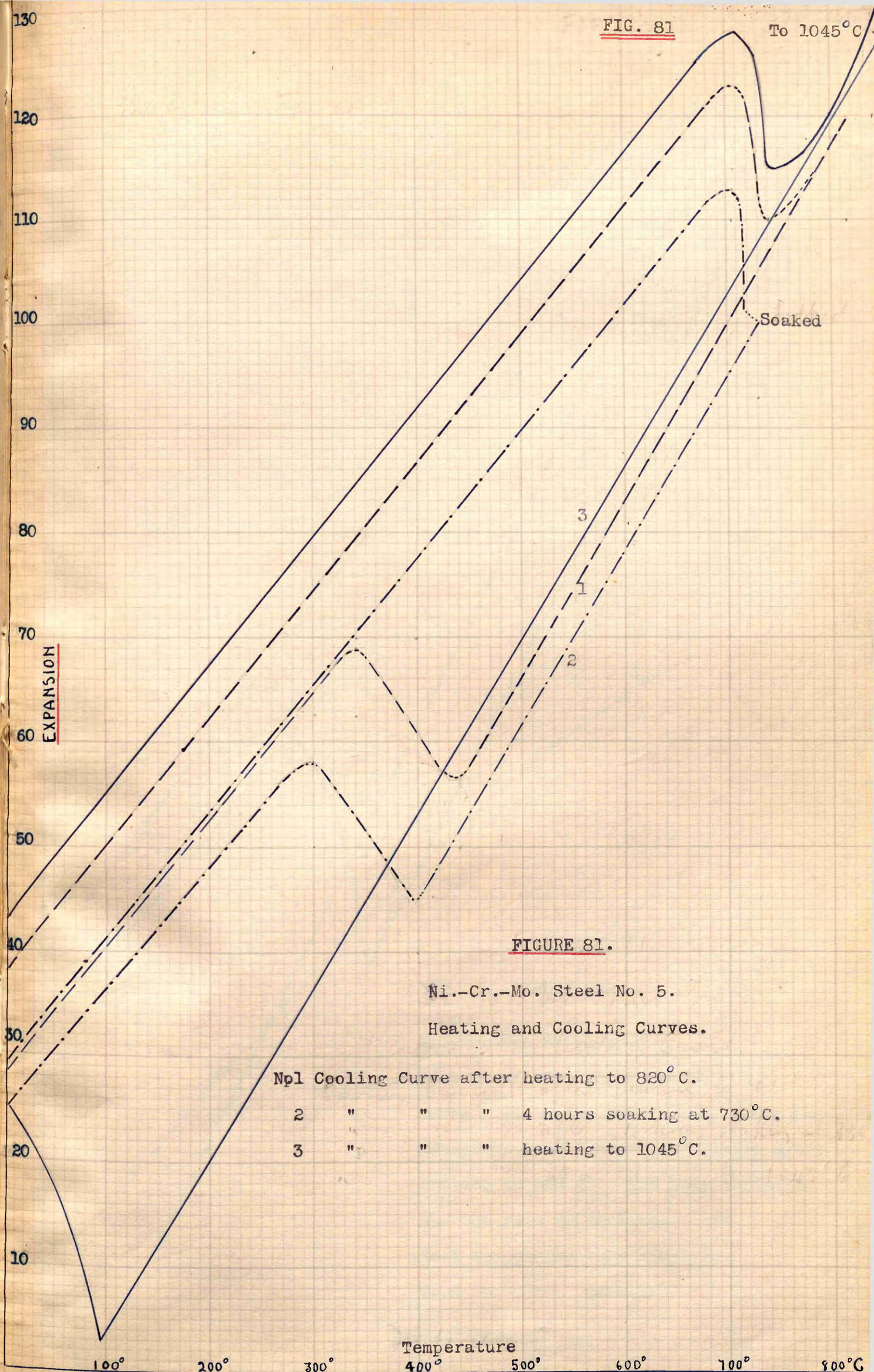


FIGURE 81.

Ni.-Cr.-Mo. Steel No. 5.

Heating and Cooling Curves.

Npl Cooling Curve after heating to 820°C.

2 " " " 4 hours soaking at 730°C.

3 " " " heating to 1045°C.

FIG. 82

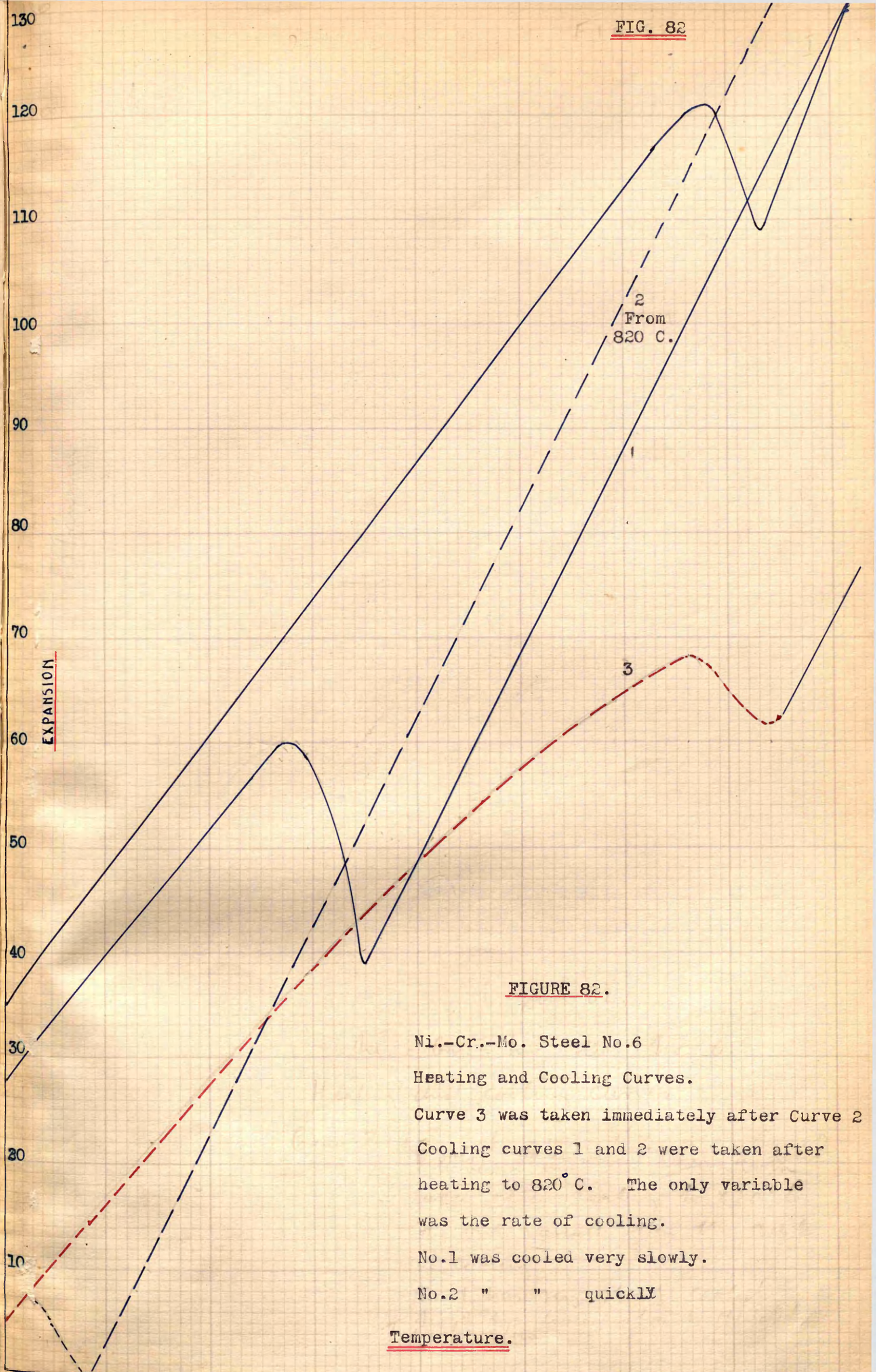


FIGURE 82.

Ni.-Cr.-Mo. Steel No.6

Heating and Cooling Curves.

Curve 3 was taken immediately after Curve 2

Cooling curves 1 and 2 were taken after heating to 820° C. The only variable was the rate of cooling.

No.1 was cooled very slowly.

No.2 " " quickly

Temperature.

FIG. 83

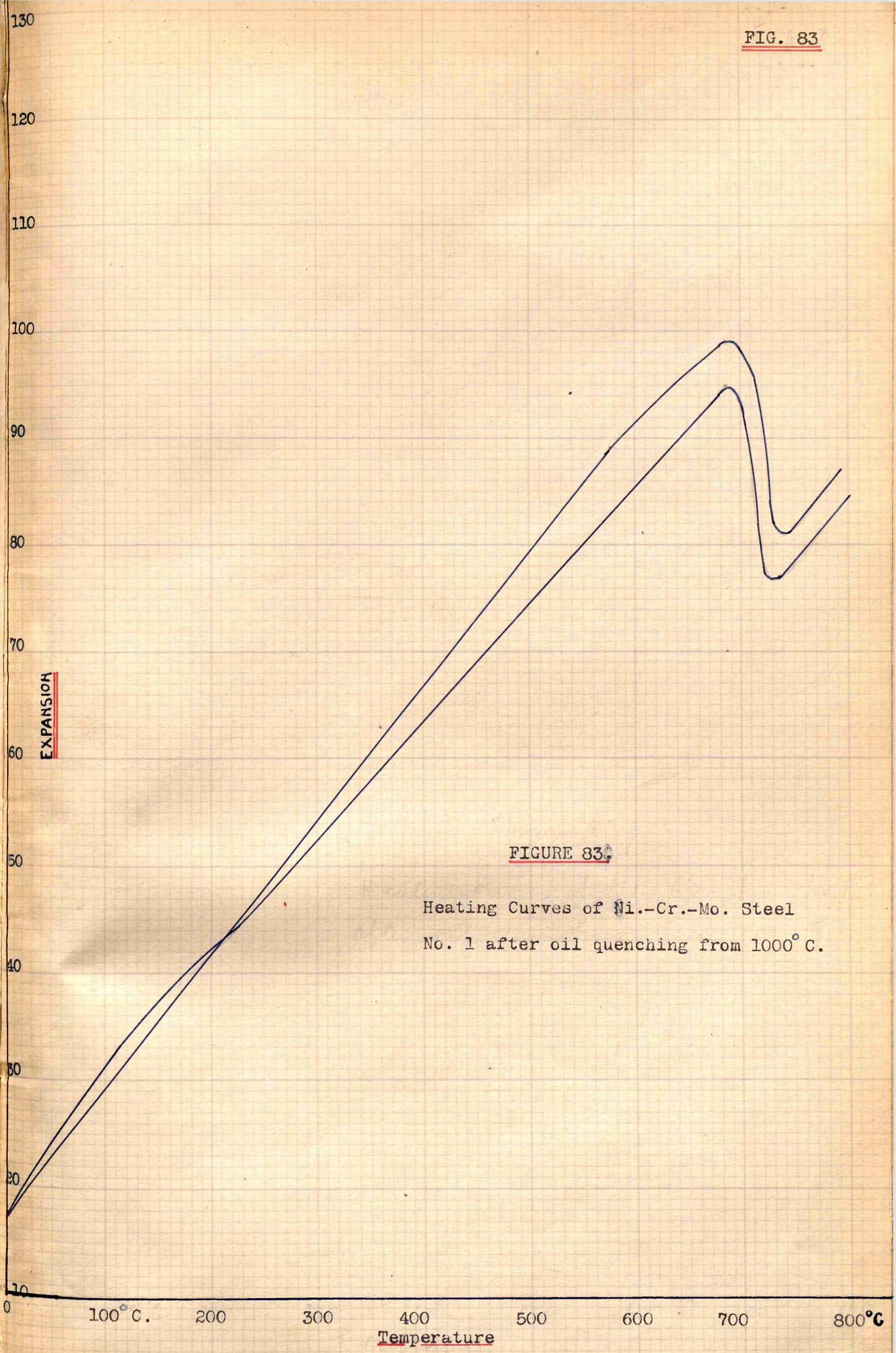


FIGURE 83.

Heating Curves of Ni.-Cr.-Mo. Steel No. 1 after oil quenching from 1000° C.

FIG. 84

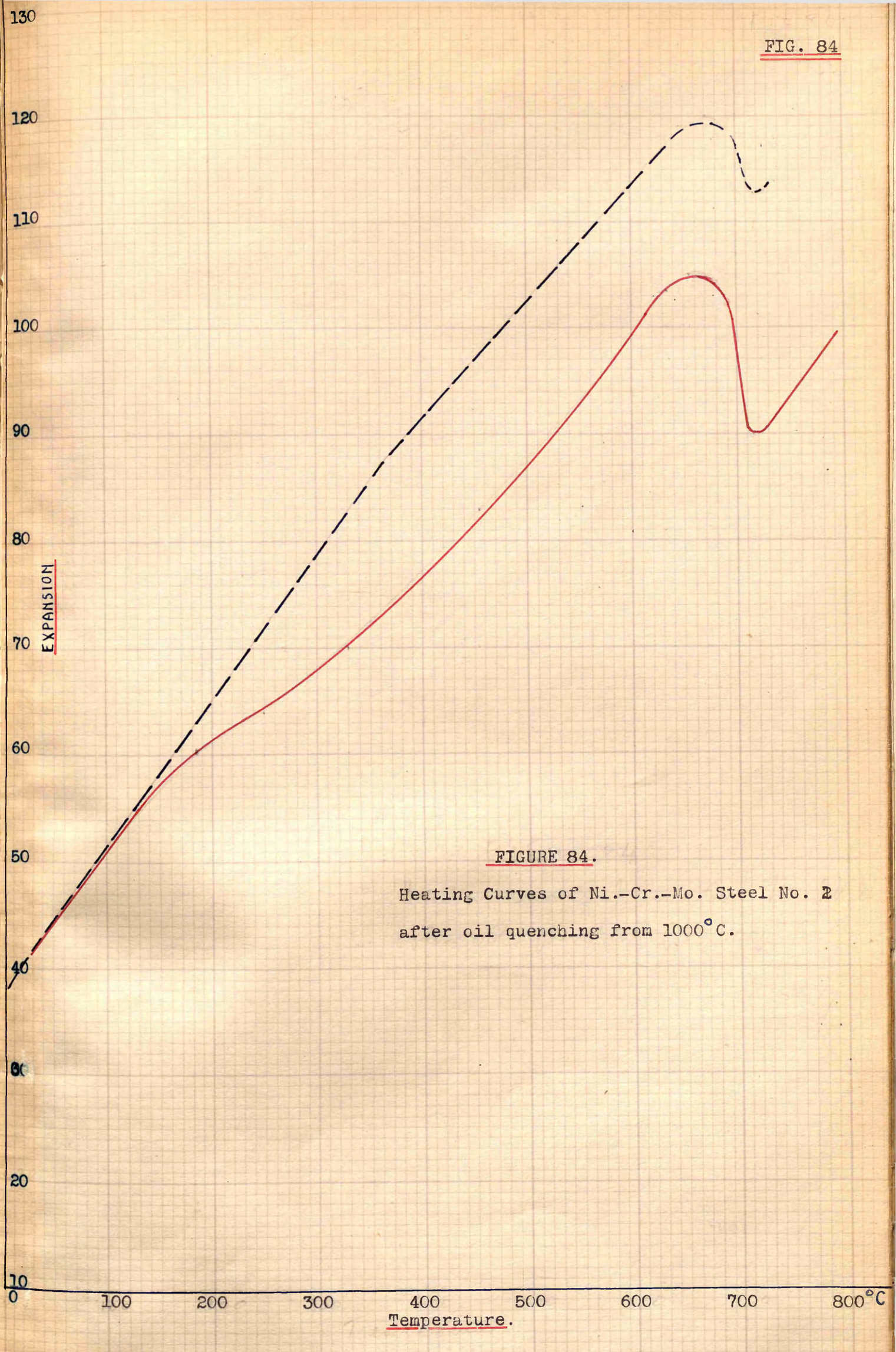


FIGURE 84.

Heating Curves of Ni.-Cr.-Mo. Steel No. 2 after oil quenching from 1000° C.

FIG. 85

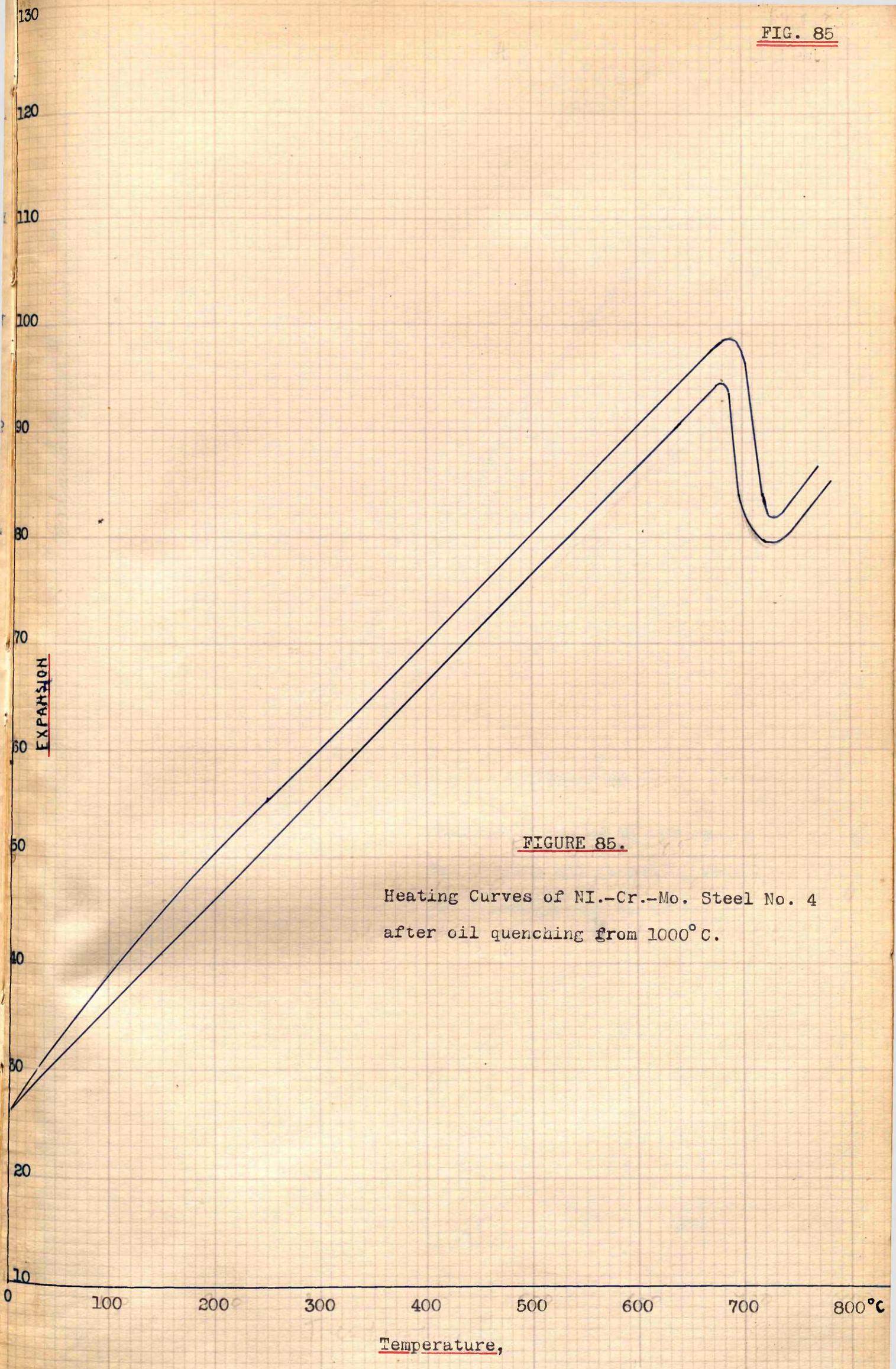


FIGURE 85.

Heating Curves of NI.-Cr.-Mo. Steel No. 4 after oil quenching from 1000° C.

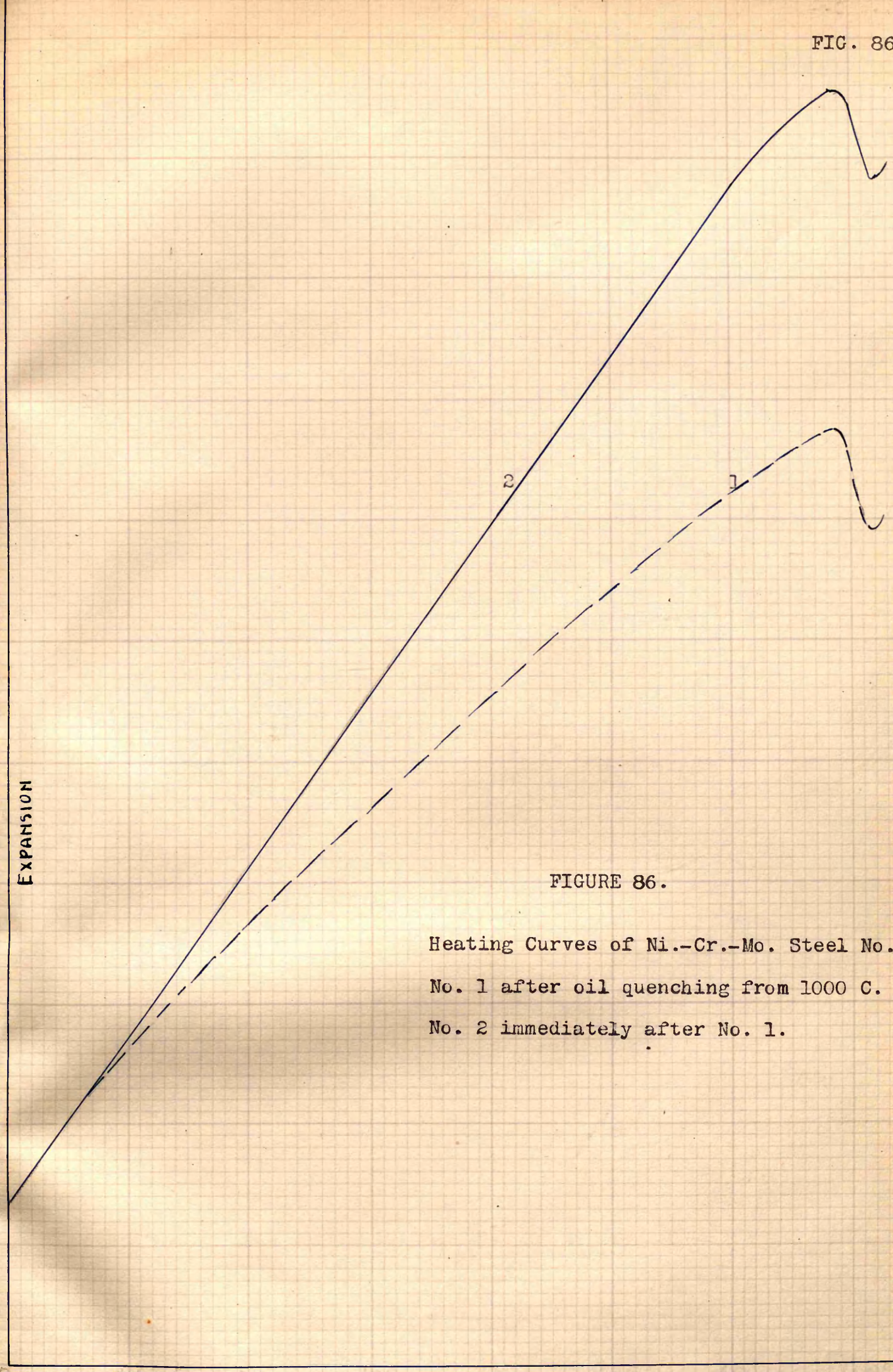


FIGURE 86.

Heating Curves of Ni.-Cr.-Mo. Steel No. 5.

No. 1 after oil quenching from 1000 C.

No. 2 immediately after No. 1.